TRANSACTIONS OF THE

AMERICAN

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FOR METALS

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INFLUENCE OF AUSTENITIC GRAIN SIZE ON THE CRITICAL COOLING RATE OF HIGH PURITY IRON-CARBON ALLOYS

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Abstract

A method is described for heating small specimens in vacuo and in an atmosphere of dry nitrogen to different temperatures and quenching directly in hydrogen. Determinations were made of the austenitic grain size and critical cooling rate of high purity iron-carbon alloys ranging in carbon from 0.23 to 1.21 per cent. These data show how carbon influences the austenitic grain size and critical cooling rate of the alloys and make possible a direct comparison of the relation of austenitic grain size to critical cooling rate or hardenability.

Introduction

THE critical cooling rate, or hardenability of a steel at the time of quenching depends primarily upon three factors, namely, (a) chemical composition, (b) chemical homogeneity and (c) grain size. The effect of carbon content on the rate of transformation of austenite in high purity iron-carbon alloys and in plain carbon steels (1),* and the influence of variations in chemical homogeneity on the mode and rate of transformation of austenite on quenching high purity iron-carbon alloys (2), were discussed in previous reports. The present study was made to determine quantitatively the influence of grain size of the austenite on the critical cooling rate of the iron-carbon alloys.

Alloys Studied

The high purity iron-carbon alloys used in the present study ranged in carbon from 0.23 to 1.21 per cent. The alloys were the same as those used in previous investigations and their preparation was described in detail in a previous report (1). Essentially, the

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*The figures appearing in parentheses refer to the references appended to this paper.

procedure for preparing these alloys consisted in carburizing hot- and cold-worked specimens of high purity iron in a mixture of hydrogen and benzene vapor, and subsequently homogenizing by heating in vacuo at 1700 degrees Fahr. (925 degrees Cent.). Cooling from the temperature used in homogenizing was sufficiently rapid to produce

Table I
Impurities Determined in the Iron-Carbon Alloys

| | Method of Analysis[1] | | |
Element	Spectro-Chemical[2] Per Cent by Weight	Chemical Per Cent by Weight	Vacuum Fusion Per Cent by Weight
Manganese	....	0.002	.....
Phosphorus	....	<0.001	.....
Sulphur	....	0.004	.....
Silicon	0.001	0.002	.....
Copper	<0.001	Not detected	.....
Nickel	0.006	0.007	.....
Cobalt	....	0.007	.....
Calcium	<0.001	.....	.....
Lead	Probably <0.001	.....	.....
Magnesium	<0.001	.....	.....
Oxygen	....	.....	0.003
Nitrogen	....	.....	0.001
Hydrogen	....	.....	0.0002

[1]Determinations were made on specimens from the bars after carburizing and heat treatment for homogeneity except as follows: Values of nickel, cobalt, and copper by chemical analysis were obtained from the electrolytic iron prior to melting and the values of manganese and phosphorus by chemical analysis were obtained from the bars as carburized. Spectro-chemical analyses were made by B. F. Scribner, chemical analyses by W. H. Jukkola and J. L. Hague, and vacuum-fusion analyses by V. C. F. Holm, all members of the staff of the National Bureau of Standards.
[2]Spectrum lines of cobalt and manganese were also found.

sorbite (fine pearlite) in all of the alloys. Specimens subsequently used for the determination of the austenitic grain size and critical cooling rate had this initial structure of sorbite. The method used in preparing the alloys eliminated the quality factor and variables other than carbon in composition.

The results of spectrochemical and chemical analyses, and the gas content as determined by the vacuum-fusion method for the iron-carbon alloys are given in Table I. Sulphur, nickel, cobalt and oxygen were the major impurities contained in the alloys. These elements amounted to about 0.021 per cent, whereas the total percentage (by weight) of all impurities determined was about 0.030.

EXPERIMENTAL PROCEDURE

All the specimens used in the present investigation were approximately 0.10 inch square by 0.040 inch thick. One wire of a 32 gage

chromel-alumel thermocouple was spot-welded to the center of one flat face of the specimen, and the other wire was spot-welded to the center of the opposite face. The specimens were heated in vacuo or in an atmosphere of dry nitrogen within a chromel coil to the desired temperature, and were held at that temperature for 15 minutes before cooling.
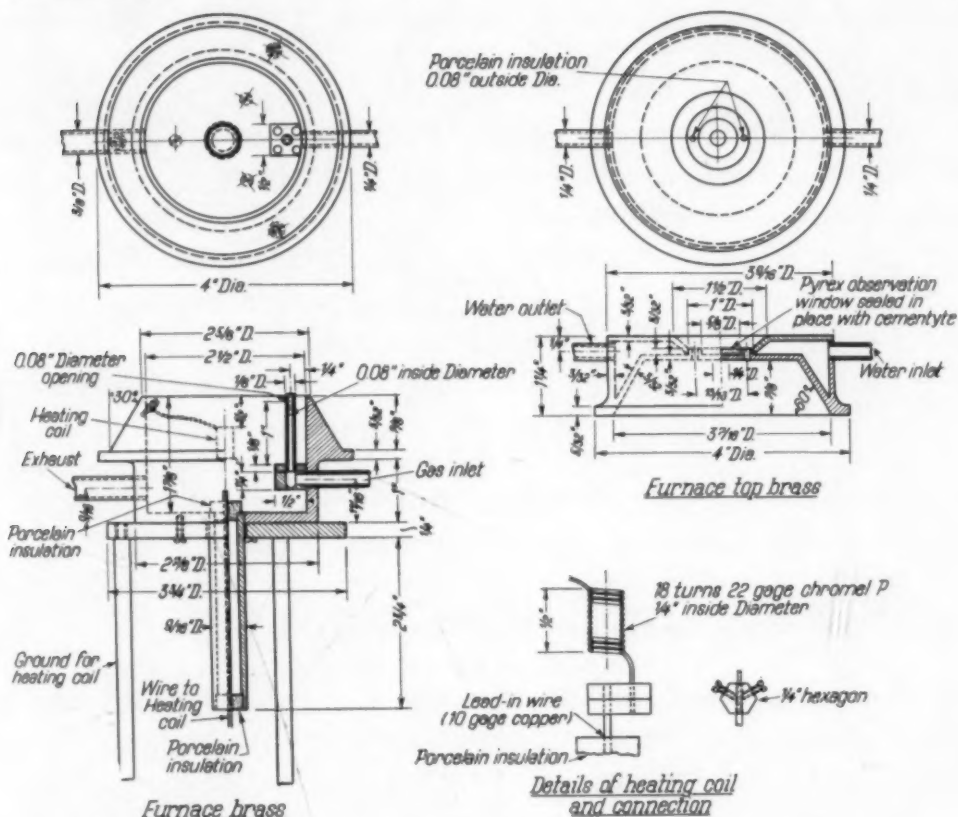
Fig. 1—Details of Furnace and Heating Coil for Heating Specimens in Vacuo or in Gas.
Test Specimen is Supported by 32-Gage Chromel-Alumel Thermocouple Wires Passing Through Porcelain Insulation in Top of Furnace and is so Adjusted as to be Located in Center of Heating Coil When Furnace is Assembled. Thermocouples and Heating Coil Wires are Sealed in Porcelain Insulation with Cementyte.

The specimens used in the determination of the austenitic grain size were cooled from the temperature establishing the grain size in a manner suitable for outlining the grains with proeutectoid ferrite or cementite, or troostite (fine pearlite). Grain size measurements were made by the method described by Jeffries (3). The specimens used in the determination of the critical cooling rate were quenched directly from the high temperatures in hydrogen. For correlating the critical cooling rate with microstructure, microscopic examination was made of the quenched specimens on the cross section of the

0.040 inch thick sheet a short distance from the point of contact of the thermocouple wires.

Fig. 1 shows the details of the construction of the furnace, and Fig. 2 shows the assembly of the apparatus used for heating the specimens either in vacuo or in dry nitrogen and quenching directly in hydrogen. The chamber (E) containing the heating coil and the specimen was evacuated before heating the specimen. For heating
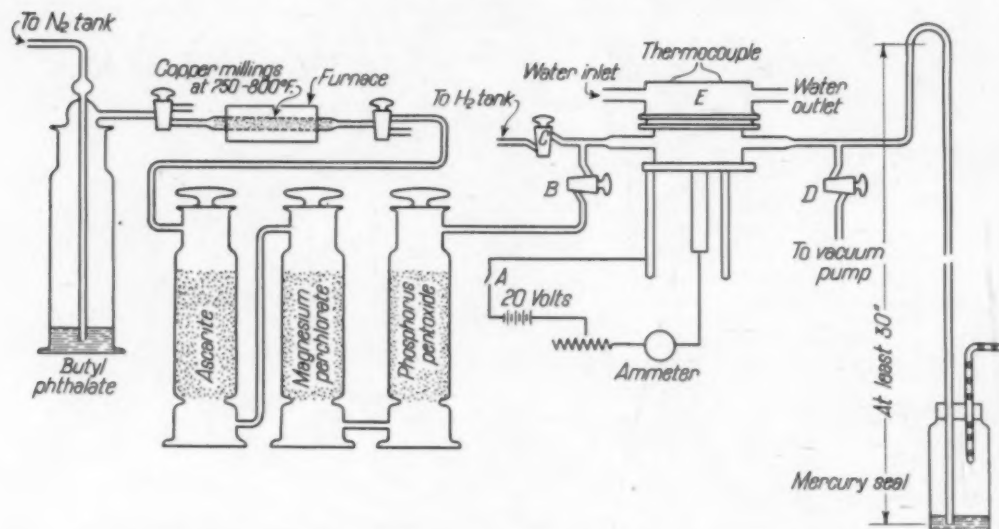
Fig. 2—Assembly of Apparatus for Heating Specimens in Vacuo or in an Atmosphere of Dry Nitrogen and Quenching Directly in Hydrogen.

the specimen in vacuo, the stop-cocks (C) to the hydrogen tank and (B) to the nitrogen train were closed, and the stop-cock (D) to the vacuum pump was opened. For heating the specimen in nitrogen, the stop-cock (C) was closed, stop-cock (D) was opened only until the chamber was evacuated, and nitrogen was admitted into the evacuated chamber until the pressure within the chamber was slightly in excess of that of the atmosphere as shown by the mercury seal. The nitrogen was bubbled through butyl-phthalate and over copper heated to 750 to 800 degrees Fahr. (400 to 425 degrees Cent.), and dried by passing through Ascarite, magnesium perchlorate, and phosphorus pentoxide. The stop-cock (B) to the nitrogen train was closed prior to heating the specimen. At the time of quenching the specimen heated either in vacuo or nitrogen, the stop-cocks (D) and (B) were closed, and switch (A) opened and the stop-cock (C) turned to permit passage of hydrogen into the heating chamber. The hydrogen after passing around the specimen, was exhausted into the air through the mercury seal. The desired cooling rates were ob-

tained by regulating the flow of gas with reducing and needle valves attached to the hydrogen tank, and by changing the position of a gas nozzle within the chamber. With a selected rate of flow of hydrogen, higher cooling rates were obtained when the outlet of the nozzle was directly below the heating coil containing the specimen, than when the nozzle was in the side position, shown in Fig. 1. After the heating chamber was partially evacuated, three "C" clamps (not shown in Figs. 1 and 2) were used to clamp together the top and bottom sections of the furnace. This procedure aided in obtaining a tight seal with the stop-cock grease used along the tapered surfaces of the furnace, and also prevented the furnace top from lifting during the time of quenching. The stop-cocks B, C and D were specially ground to prevent leakage, and a liquid air trap was used between the furnace and the mercury diffusion pump.

With the present equipment, and with the use of gases as the quenching media it was possible to heat the specimens in vacuo or dry nitrogen at various rates to different temperatures and to quench directly from these temperatures without moving the specimens. Oxidation of the specimens during the time of heating and quenching was eliminated, and progressive changes in cooling rates were obtainable over the wide range required in the present experiments. The structure produced in the quenched specimens indicated that they were cooled uniformly over the entire surface.

Preliminary tests were carried out to determine whether specimens of high purity iron were contaminated with aluminum, chromium, and nickel in the process of spot-welding the chromel-alumel thermocouple, or during the time of heating the specimens to high temperatures in contact with this couple within a chromel coil in vacuo. Spectrochemical analysis made on these specimens after removing 0.002 and 0.003 inch from the original surface showed no indication of aluminum or chromium, and only a small increase in nickel incident to these treatments.

For the determination of the critical cooling rates, a photographic time-temperature cooling curve was obtained during the quench by means of a string galvanometer apparatus, in the manner previously described by French and Klopsch (4). The string galvanometer used consisted of a fine tungsten wire suspended in a strong magnetic field. When this galvanometer is connected to a source of electromotive force, which in this case is a thermocouple, a deflection of the string is obtained which is proportional to the current in the string

circuit. Provided the resistance of the string circuit and the strength of the magnetic field remain constant, deflections of the string are directly proportional to the electromotive force and these deflections may be converted into temperature equivalents. The shadow of the string is projected, at about 500 magnification, upon a moving photographic film of bromide paper, 2⅜ inches wide, thus giving a continuous record of variations in temperature during the time of the
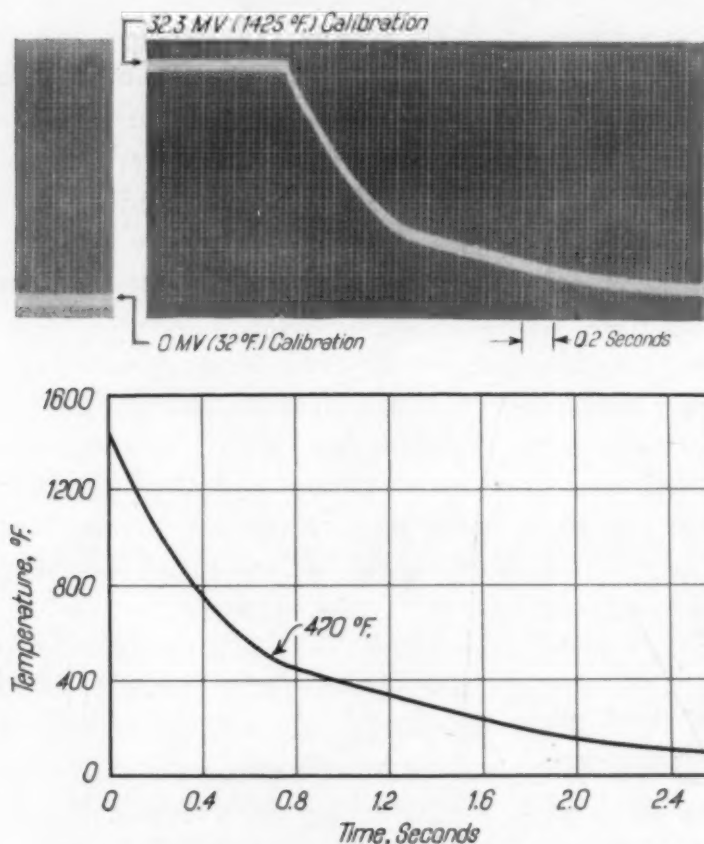
Fig. 3—Photographic Record Obtained with the String Gal-vanometer Apparatus and Plotted Time-Temperature Cooling Curve. Specimen of the Alloy Containing 0.85 Per Cent Carbon Heated in an Atmosphere of Dry Nitrogen and Quenched Directly in Hydrogen at About its Critical Cooling Rate.

quench. Time intervals are recorded on the film by interrupting the beam of light with a slotted disk rotated at a constant speed by means of a synchronous motor operated from a tuning fork. A continuous time-temperature cooling curve could be plotted from the data recorded on the film.

Calibration of this apparatus was carried out before each quench by recording on the film the zero position of the string, corresponding

to open circuit of the thermocouple (zero e.m.f.), and the position of the string when the known e.m.f. of the thermocouple corresponding to the quenching temperature was applied (Figs. 3 and 4). The e.m.f. of the thermocouple corresponding to the quenching temperature was determined with a potentiometer indicator. Change in resistance of the galvanometer circuit resulting from cooling a short length of the thermocouple wires during the time of quenching was
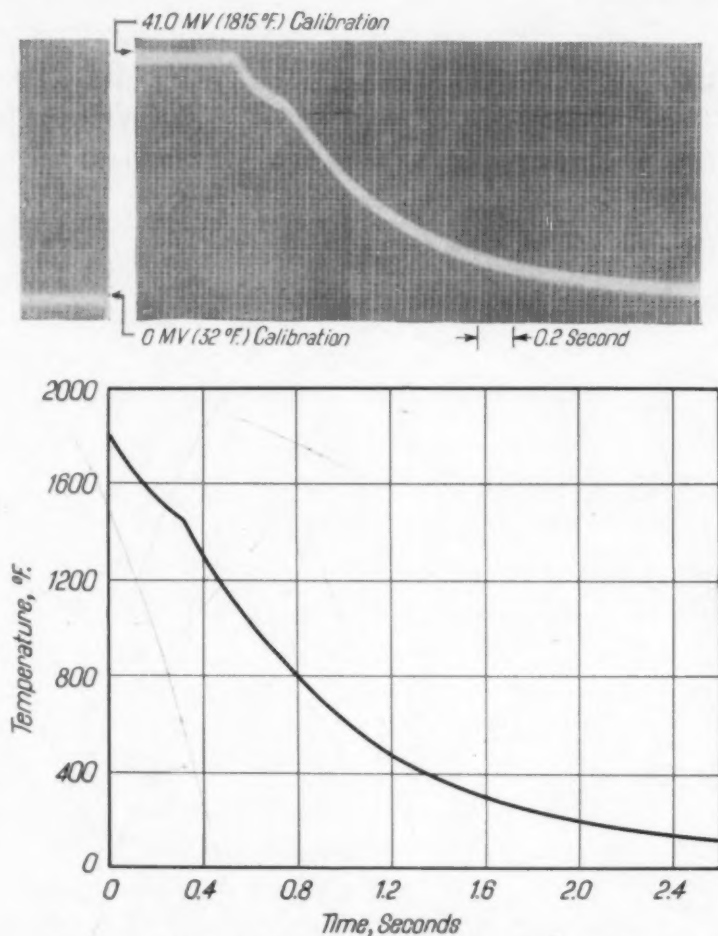
Fig. 4—Photographic Record Obtained with the String Galvanometer Apparatus and Plotted Time-Temperature Cooling Curve. Platinum Specimen Heated in Vacuo and Quenched Directly in Hydrogen.

relatively small in comparison with the total resistance of the circuit. This variation in resistance was well within the limit of experimental error and could be neglected.

Fig. 3 shows a photographic record and plot of a time-temperature cooling curve obtained on a specimen of the 0.85 per cent carbon alloy heated in an atmosphere of dry nitrogen to 1425 degrees Fahr.

(775 degrees Cent.) and quenched directly in hydrogen at about its critical cooling rate. No arrest appeared in the cooling curve until the temperature of the specimen had fallen to about 470 degrees Fahr. (245 degrees Cent.). At approximately this temperature the austenite started to transform to martensite (Ar″). The Ar″ transformation, however, was not clearly shown in the photographic records obtained on quenching the relatively stable austenite of the hypereutectoid alloys. With the hypereutectoid alloys, the Ar″ occurred at temperatures below the arrest shown in Fig. 3, a factor that made difficult its detection.

A photographic record and a plot of the time-temperature curve obtained on a platinum specimen heated in vacuo to 1800 degrees Fahr. (980 degrees Cent.) and quenched directly in hydrogen are shown in Fig. 4. There was an arrest in the cooling curve at about 1450 degrees Fahr. (785 degrees Cent.), or about 350 degrees Fahr. (195 degrees Cent.) below the temperature at the start of the quench. The arrest, obviously, was not due to a heat evolution caused by a structural change in the platinum specimen. This high temperature arrest was observed in the cooling curves of all the specimens heated in vacuo and rapidly quenched directly in hydrogen, but it was not observed in the specimens heated in an atmosphere of nitrogen and similarly quenched. The arrest was therefore characteristic of the procedure used in heating in vacuo and quenching directly in hydrogen and was probably due to the variation in rate of flow of the quenching gas around the specimen during a short interval at the start of the quench. Such an arrest in the cooling of specimens by heating in vacuo and quenching directly in hydrogen has no appreciable effect on the rate of transformation of austenite provided the quenching temperature is sufficiently high for the lag to occur above or in the vicinity of Ae_1. With quenching temperatures just above Ae_1, the arrest may occur in the upper range in which austenite is least stable (1110 to 930 degrees Fahr.) and thus have an influence in the formation of the decomposition products of austenite of high transformation rate. To eliminate this factor in the determination of the critical cooling rate, the specimens quenched directly from temperatures just above Ae_1 were heated in an atmosphere of dry nitrogen.

As shown in Fig. 4, no arrest occurred on cooling the platinum specimen through the temperature range below 950 degrees Fahr. (510 degrees Cent.). With high purity iron-carbon alloys containing

Fig. 5—Effect of Rate of Heating Through the Transformation Range on the Grain Size at 1425 Degrees Fahr. of Iron-Carbon Alloy Containing 0.73 Per Cent Carbon.

A—Heated Rapidly in Lead Bath at 1425 Degrees Fahr. Approximately 40 Grains Per Square Inch at 100 Diameters. B—Heated from 70 to 1400 Degrees Fahr. in 1 Minute. Approximately 20 Grains Per Square Inch at 100 Diameters. C—Heated from 70 to 1415 Degrees Fahr. in 2½ Minutes. Approximately 5 Grains Per Square Inch at 100 Diameters. Etched with 1 Per Cent Nital. × 100.
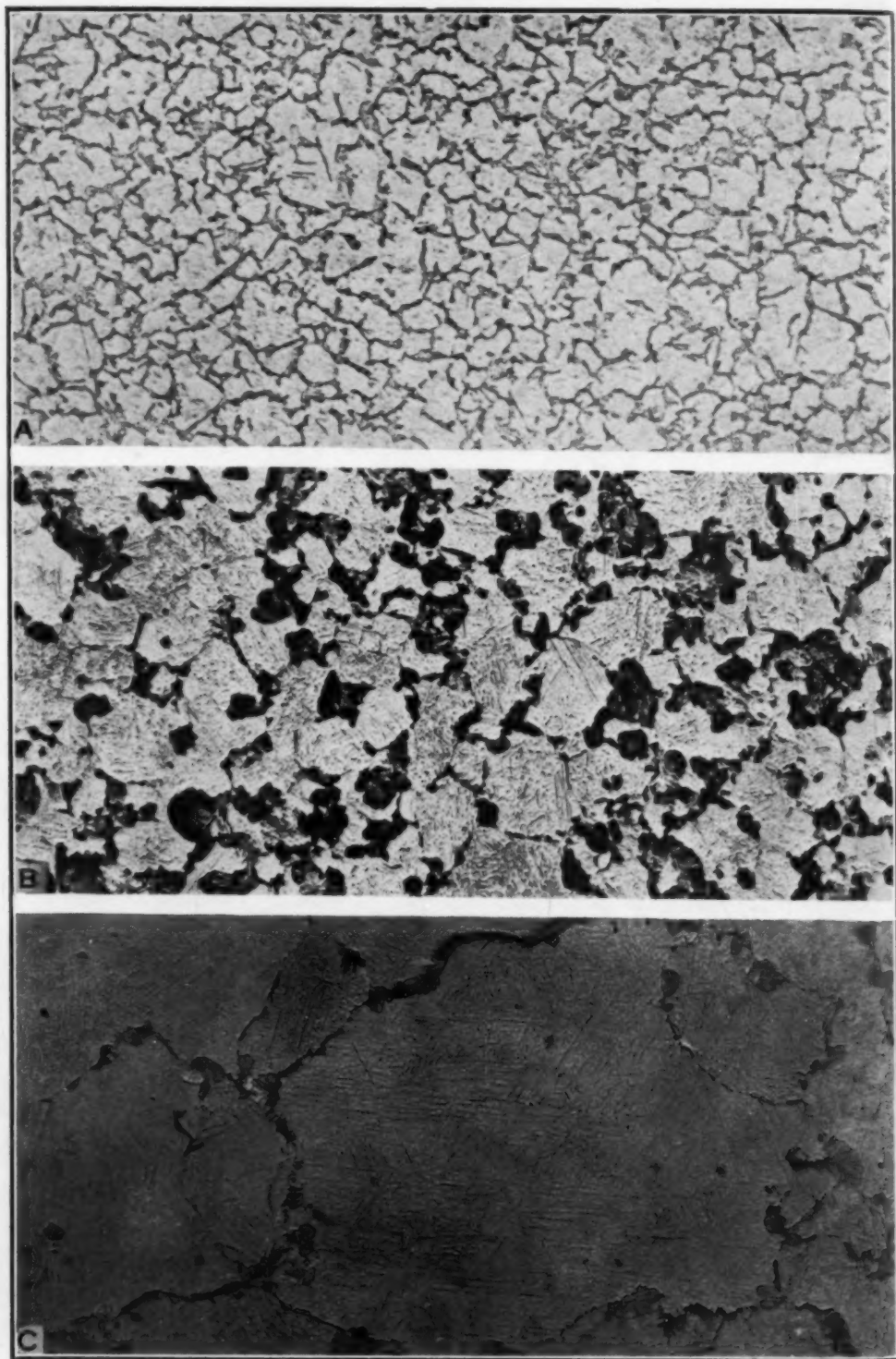
Fig. 6—Effect of Rate of Heating Through the Transformation Range on the Grain Size at 1600 Degrees Fahr. of Iron-Carbon Alloy Containing 0.50 Per Cent Carbon. A—Heated Rapidly in Lead Bath at 1600 Degrees Fahr. Approximately 50 Grains Per Square Inch at 100 Diameters. B—Heated from 70 to 1590 Degrees Fahr. in 51 Seconds. Approximately 15 Grains Per Square Inch at 100 Diameters. C—Heated from 70 to 1415 Degrees Fahr. in 2½ Minutes. Approximately 1 Grain Per Square Inch at 100 Diameters. Etched with 1 Per Cent Nital. × 100.

carbon in excess of 0.20 per cent, the transformation of austenite to martensite begins at temperatures below 950 degrees Fahr. (510 degrees Cent.).

Austenitic Grain Size

Previous Investigations

Reviews of some of the most important works on the subject of austenitic grain size have recently been prepared by Shapiro (5) and by Ward and Dorn (6). Shapiro states that "The final austenitic grain size of any steel is influenced by: (a) prior structure or initial grain size, (b) mechanical deformation (hot and cold work) which remains or is conferred upon it during austenitizing, (c) the rate of heating to the austenitic condition, (d) time and temperature of heating (T^{max}), (e) method of deoxidation, and (f) the type of inhibitors resulting from the deoxidization process. These factors may either intensify or minimize one another's effects and thereby may either refine or increase the austenitic grain size, depending on whether they act in a like or opposite manner."

Derge, Kommel, and Mehl (7) have investigated some of the factors influencing austenitic grain size in high purity steels. They conclude that "Al_2O_3 and SiO_2 inhibit grain growth; whereas aluminum and silicon in solid solution do not, and that the degree of inhibition is closely related to the type of dispersion of the oxides." Grossmann (8) has discussed the mechanism of grain growth in low carbon steels during carburization. He shows that upon heating a low carbon steel, the first austenite appears at the interfaces of the ferrite and carbide. As the temperature of heating is increased, the austenite begins to encroach on the ferrite grains and proceeds until the steel is completely transformed to austenite.

Effect of Rate of Heating Through the Transformation Range

Specimens of the high purity iron-carbon alloys were heated at various rates through the transformation range to temperatures ranging from 1425 to 1800 degrees Fahr. (775 to 980 degrees Cent.). The grain sizes obtained with the different rates of heating for the 0.73 per cent carbon alloy at 1425 degrees Fahr. (775 degrees Cent.) are shown in Fig. 5, for the 0.50 per cent carbon alloy at 1600 degrees Fahr. (870 degrees Cent.) in Fig. 6, and for the 1.21 per cent carbon alloy at 1800 degrees Fahr. (980 degrees Cent.), in Fig. 7. The
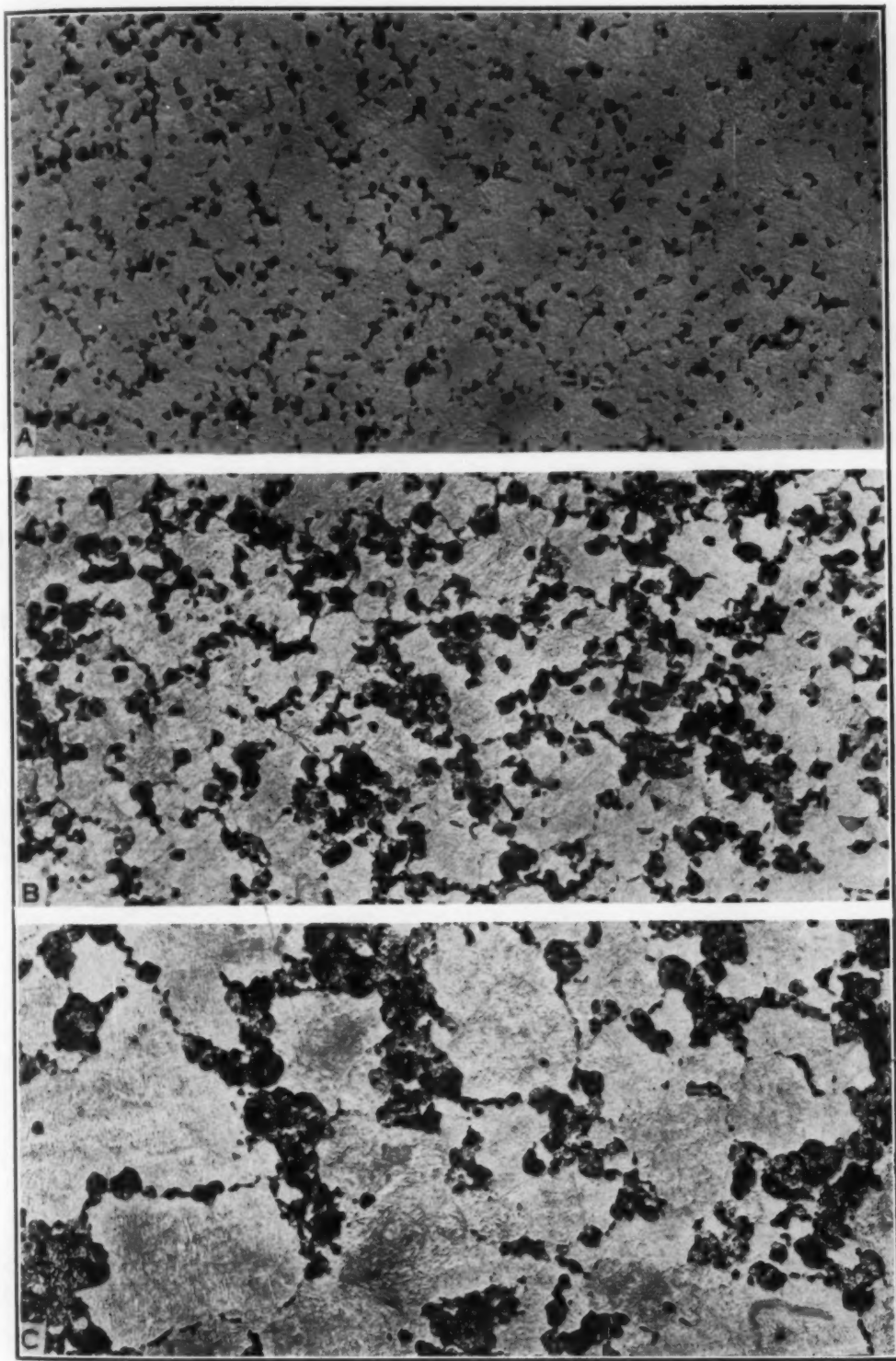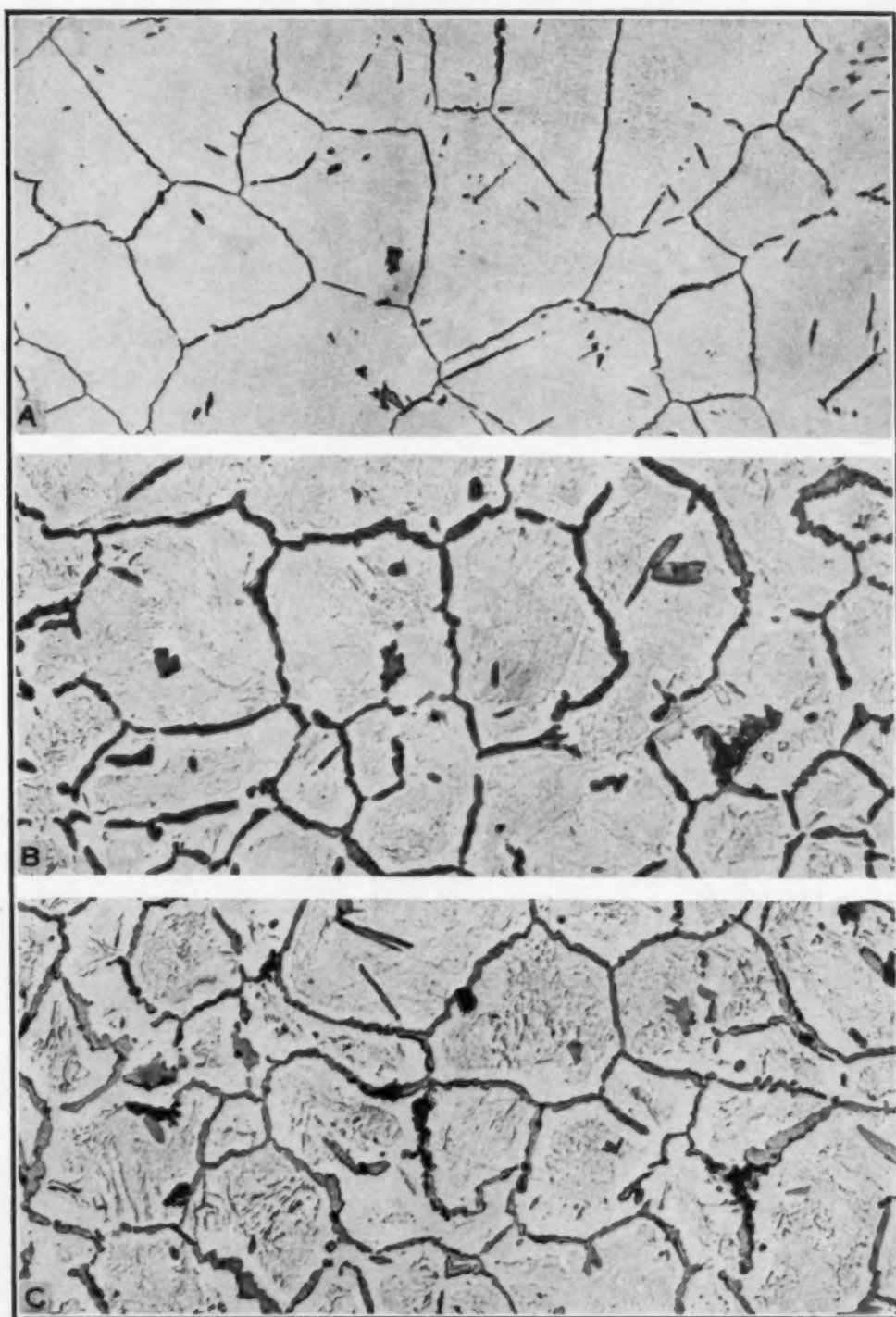
Fig. 7—Effect of Rate of Heating Through the Transformation Range on the Grain Size at 1800 Degrees Fahr. of Iron-Carbon Alloy Containing 1.21 Per Cent Carbon.
A—Heated from 70 to 1750 Degrees Fahr. in 1 Minute. Approximately 2.2 Grains Per Square Inch at 100 Diameters. B—Heated from 1330 to 1370 Degrees Fahr. in 24 Minutes. Approximately 2.3 Grains Per Square Inch at 100 Diameters. C—Heated from 1330 to 1370 Degrees Fahr. in 38½ Minutes. Approximately 3.2 Grains Per Square Inch at 100 Diameters.
Etched with Boiling Sodium Picrate. × 100.

grain size established at temperatures ranging from 1425 to 1600 degrees Fahr. (775 to 870 degrees Cent.) increased markedly with decrease in rate of heating these high purity alloys, whereas the grain size at 1800 degrees Fahr. (980 degrees Cent.) was not so noticeably dependent upon the rate of heating. Additional data on this subject will be presented in a subsequent report.

Table II
Austenitic Grain Size of the Iron-Carbon Alloys

The specimens 0.1 inch square by 0.040 inch thick were heated rapidly in vacuo or in an atmosphere of dry nitrogen to temperatures indicated, held at temperature for 15 minutes and then cooled in a manner suitable for outlining the grains. In general, each grain size value reported is the average of four determinations made on one specimen.

Carbon Per Cent	Average Number of Grains Per Square Inch at 100 Diameters at Various Temperatures, °F.											
	1425	1500	1550	1600	1650	1700	1750	1800	1900	2000	2050	2100
0.23	..	....	..	....	..	(8)[2]	..	2.2, 2.2	...	1.1	...	0.8
0.40	..	....	..	....	..	....	..	2.3, 2.7	...	...	...	...
0.50	..	12	..	12, 9 7	..	....	..	2.3, 3.0	...	...	...	...
0.62	13, 15	10, 8	..	9	..	8, (8), (8)	..	2.2, 3.2, 3.7, 3.5	...	...	...	1.5
0.73	23	17	..	13, 12	..	....	..	2.6, 2.5, 2.7 3.5, 3.5, 3.0	...	...	1.6	...
0.80	..	25	14	11	...	....	..	3.0, 2.7	...	...	...	...
0.85	..	17	..	....	..	8	..	2.8	...	0.8	...	...
1.01	..	....	..	16, 15 10	..	6, (8)	..	3.1	2.1	1.6	...	...
1.14	..	....	..	....	9	6, 7, (9), (8)	4	3.1	2.1	1.9	...	...
1.21	..	....	..	....	..	9, (9)	..	2.8, 1.9, 2.3, 2.7, 3.5, 3.1, 3.7	...	...	...	...

[2]The values appearing in parentheses were obtained in a previous investigation (1) on specimens ¼ inch square by 0.040 inch thick, heated rapidly in dry nitrogen.

Effect of Temperature

The grain sizes at temperatures ranging from 1425 to 2100 degrees Fahr. (775 to 1150 degrees Cent.) were determined for the alloys and the results are summarized in Table II. Although the specimens were heated at a rapid rate through the transformation range, this rate was not precisely the same for all the specimens. With the heating rate employed and with complete solution of carbon, all the alloys had approximately the same average grain size at any selected temperature within the range of 1600 to 2100 degrees Fahr. (870 to 1150 degrees Cent.). At temperatures of 1425 and 1500 degrees Fahr. (775 and 815 degrees Cent.), some variations were observed in the average grain size of the different alloys which were probably due to small differences in rate of heating or to masking of

some of the smaller size grains by coalescence of the constituent used in outlining the parent austenite grain.

Carbon was the only element present in these alloys in a sufficient amount that might be considered in the class of grain growth inhibitors. At temperatures required for complete solution in austenite (that is above Ac_3 or Ac_{cm}), carbon evidently is not effective in inhibiting grain growth in high purity iron-carbon alloys. For example, at 1700 degrees Fahr. (925 degrees Cent.), each of the alloys ranging in carbon from 0.23 to 1.21 per cent had an average grain size of about 6 to 9 grains per square inch at 100 diameters. This temperature is only slightly above the Ac_{cm} for the 1.21 per cent carbon alloy, whereas it is approximately 370 degrees Fahr. (205 degrees Cent.) in excess of the temperature required for complete solution of carbon in the alloy of eutectoid composition. The actual temperature (T^{max}) and the rate of heating through the transformation range were the dominant factors in controlling the austenitic grain size of these alloys.

CRITICAL COOLING RATE

Davenport and Bain (9), and co-workers (10), in their studies of the process and results of the transformation of austenite at constant temperatures, showed that the structure and hardness of a quenched steel depends primarily on the particular temperature at which the austenite decomposes during quenching. The general form of the transformation temperature-time curve (the so-called S-curve when temperature is plotted on Cartesian and time on a logarithmic basis) was the same for all steels investigated, but there were significant differences in the position of the S-curve on the time scale and minor differences on the temperature scale. Their results showed that the hardenability of a steel depends upon the stability of the quenched austenite in the temperature range of about 1110 to 930 degrees Fahr. (600 to 500 degrees Cent.), the upper bend of the S-curve, Fig. 8. This is the upper range in which austenite is most likely to decompose and in which its decomposition products are relatively soft. If by a continuous quench to room temperature, the austenite is cooled through this temperature range without decomposing (Fig. 8, curve A), it transforms only in the temperature range below about 300 degrees Fahr. (150 degrees Cent.) and the product of decomposition is martensite. If the austenite is cooled at

a slower rate through this upper temperature range (curve C), part
of the austenite transforms in this temperature range to troostite and
the remaining austenite is cooled unchanged to the temperature favor-
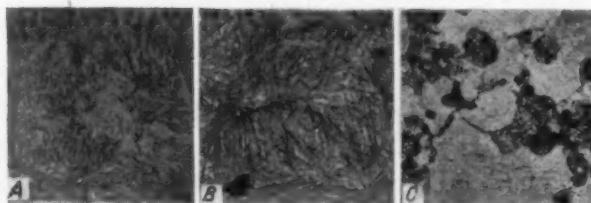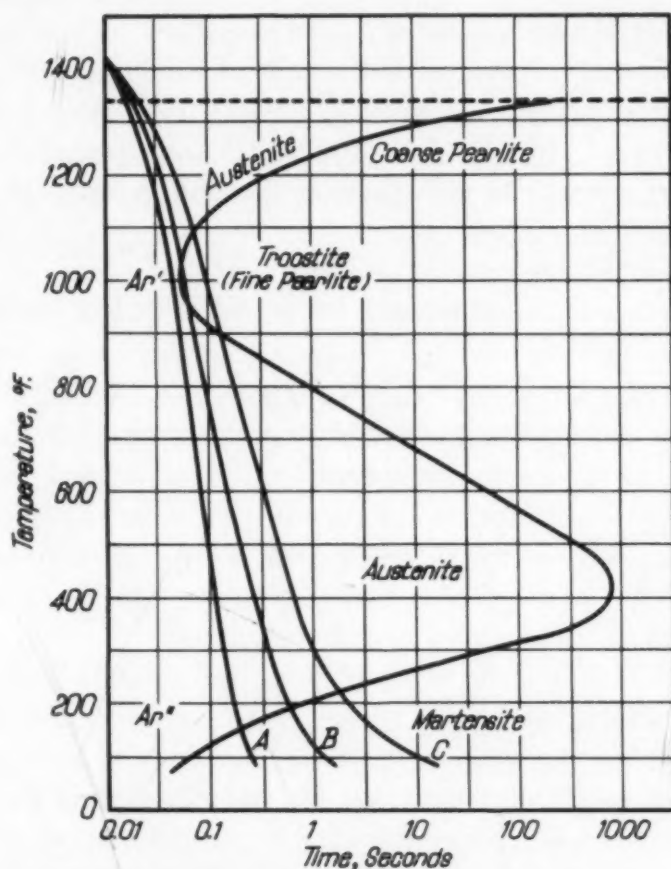
Fig. 8—Decomposition of Austenite at Various Tem-
peratures.
The Diagram Illustrates the Method of Determining
the Critical Cooling Rate and the Structures Obtained with
Different Cooling Rates.
A—Martensite.   × 250.   B—Martensite with Small
Amount of Troostite (Critical Cooling Rate).   × 250.   C—
Martensite and Troostite.   × 100.   All Specimens Etched
with 1 Per Cent Nital.

able for its decomposition to martensite.   At some cooling rate, called
the critical cooling rate, the austenite just begins to decompose (curve
B), in the range 1110 to 930 degrees Fahr. (600 to 500 degrees

Cent.). Obviously, the critical cooling rate of a steel is an index to the stability of its austenite in this temperature range. The slower the critical cooling rate the more stable is its austenite in the temperature range 1110 to 930 degrees Fahr. (600 to 500 degrees Cent.) and the deeper is the hardening.

In the present experiments, the critical cooling rate was taken as the average cooling rate between 1110 and 930 degrees Fahr. (600 and 500 degrees Cent.), which produced in the quenched specimens a structure of martensite with nodular troostite in amounts estimated to be between 1 and 3 per cent.

As shown in previous experiments (2) in which specimens of the alloys were quenched directly from 1800 degrees Fahr. (980 degrees Cent.), the Ar″ transformation commenced in some of the alloys at temperatures appreciably higher than in the plain carbon steels used in the interrupted quench experiments of Davenport and Bain. Data on the isothermal transformation of austenite in high purity iron-carbon alloys would be of particular assistance in the determination of the critical cooling rates by definitely establishing the upper range in which the austenite is most likely to decompose. Accurate data on this subject would be difficult to obtain because of the relatively high transformation rates of austenite in high purity iron-carbon alloys, especially those of the lower carbon contents.

Although the transformation rates of austenite at constant temperature levels were not determined for the alloys, some data obtained in the continuous quench experiments throw light upon the temperature of the Ar′ and Ar″ transformations. These data for the alloys, ranging in carbon from 0.62 to 1.01 per cent, are summarized in Table III and in Figs. 9 and 10. In general, the start of the Ar′ occurred within the temperature range of about 1110 to 930 degrees Fahr. (600 to 500 degrees Cent.), provided the specimens were cooled at the most rapid rate possible for the austenite to transform completely to troostite. This fact is shown by the cooling curve and microstructure, reproduced in Fig. 9, for a specimen of the 0.80 per cent carbon alloy quenched directly from 1425 degrees Fahr. (775 degrees Cent.). The austenite was cooled unchanged to a temperature of about 1090 degrees Fahr. (585 degrees Cent.), which is in the upper portion of the range 1110 to 930 degrees Fahr. (600 to 500 degrees Cent.). At this temperature the austenite started to decompose as shown by the inflection in the cooling curve, the evolution of heat accompanying the transformation causing a 45 degrees

Fahr. rise in temperature to a maximum of 1135 degrees Fahr. (615 degrees Cent.) in 0.2 second. The transformation of the austenite continued at approximately this temperature for about 0.1 second and additional transformation of austenite to troostite undoubtedly occurred while the specimen cooled through the Ar′ temperature range.

A small proportion of the austenite was cooled to the temperature (Ar″) favorable for its transformation to martensite. This is shown by the microstructure of the quenched specimen, which con-

Table III
Temperature of the Start of the Ar′ and Ar″ Transformations in the Iron-Carbon Alloys

Carbon Per Cent	Quenching Temp. °F.	Transformation, °F.			Structure
		Ar′ Start	Maximum	Ar″ Start	
0.62	1800	1130	1130	...	Troostite + martensite
0.73	1425	1130	1155	...	Troostite + trace martensite
0.73	1425	1075	1100	...	Troostite + trace martensite
0.73	1425	1020	1020	...	Troostite + martensite
0.73	1800	1075	1075	...	Troostite + martensite
0.80	1425	1040	1060	...	Troostite + martensite
0.80	1425	1040	1040	500	Troostite + martensite
0.80	1425	1090	1135	...	Troostite + trace martensite
0.80	1500	1120	1190	...	Troostite
0.80	1500	1075	1075	480	Troostite + martensite
0.80	1550	1020	1020	...	Troostite + martensite
0.85	1425	1030	1030	...	Troostite + martensite
0.85	1800	1120	1120	...	Troostite + martensite
1.01	1425	1105	1210	...	Troostite + free carbides

sisted of a trace of martensite in a matrix of troostite. If the cooling rate of the alloy were progressively increased, the proportion of the austenite that transforms in the Ar′ temperature range would be decreased, the heat effect from the Ar′ transformation would be diminished in intensity and the temperature of the start of Ar′ would be lowered. With a cooling rate that produced 5 to 10 per cent troostite in the 0.80 per cent carbon alloy, the Ar′ started at approximately 1040 degrees Fahr. (560 degrees Cent.). With rates in excess of the critical cooling rate, all the austenite was cooled unchanged through the Ar′ to the Ar″ temperature range.

As already pointed out (2), the temperature at the start of the Ar″ transformation progressively decreased with increase in the carbon content from about 900 degrees Fahr. (480 degrees Cent.) for the 0.23 per cent carbon, to about 500 degrees Fahr. (260 degrees Cent.) for the 0.80 per cent carbon alloy. Increase in quenching temperature of the 0.80 per cent carbon alloy from 1425 to 1800 degrees Fahr. (775 to 980 degrees Cent.), which resulted in an increase

in austenitic grain size from approximately 20 to 3 grains per square inch at 100 diameters, had no appreciable effect on the temperature of the start of the Ar″. As shown in Fig. 10, variation in observed cooling rates from about 550 to 2600 degrees Fahr. (305 to 1445 degrees Cent.) per second had no appreciable influence on the temperature range of the start of the Ar″ transformation in the 0.80 per cent carbon alloy.
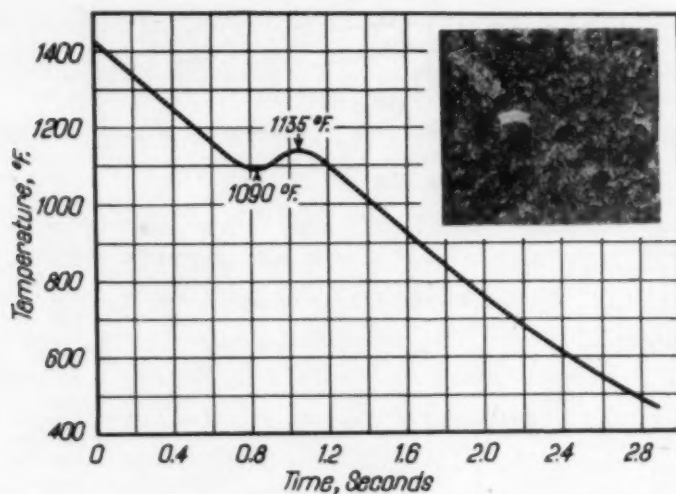
Fig. 9—Time-Temperature Cooling Curve and Structure of Iron-Carbon Alloy Containing 0.80 Per Cent Carbon. Quenched Structure Consisted of Troostite with a Trace of Martensite. Etched with 1 Per Cent Nital. ✕. 100.

The difficulty of accurate measurements of the isothermal transformation of austenite in the alloys is well illustrated by the cooling curve of Fig. 9. The specimen, containing 0.80 per cent carbon, was cooled from the quenching temperature of 1425 to below 1000 degrees Fahr. (775 to 540 degrees Cent.) in less than 1.5 seconds. This elapsed time was sufficient for the austenite to decompose to troostite. With austenite of lower carbon content, or smaller grain size, the reaction time is less than that shown for the 0.80 per cent carbon alloy.

The position of the bend in the S-curve (at 1110 to 930 degrees Fahr.) for the alloys is not accurately defined by these data (Table III, Figs. 9 and 10). Although desirable, it is not entirely essential that the location of this temperature range be known for use in the determination of the critical cooling rate of the alloys because of the exponential type of cooling curve obtained by quenching the specimens in a gas. In any event, the relation of the critical cooling rate

to the several variables studied would have been similar if some other temperature interval within the limits of approximately 1150 to 900 degrees Fahr. (620 to 480 degrees Cent.) had been selected for the determination of the critical cooling rate of the alloys.

The influence of variations in quenching rate on the rate of
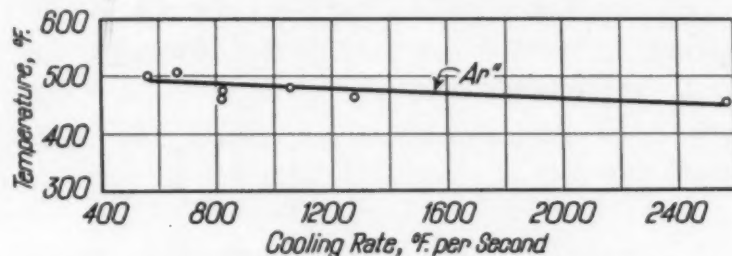
Fig. 10—Relation of Cooling Rate to Temperature of the Start of Ar″ for Iron-Carbon Alloy Containing 0.80 Per Cent Carbon Quenched from 1425 Degrees Fahr. Except for the Cooling Rate of 560 Degrees Fahr. per Second, Where a Correction was Made for the Heat Effect Due to the Ar′ Transformation, All Cooling Rates are Based on the Average Rate Required for the Specimens to Cool Through the Temperature Range of 1110 to 930 Degrees Fahr.

transformation of austenite, and on the hardness of the alloy containing 0.80 per cent carbon, are summarized in Fig. 11. At the quenching temperature of 1425 degrees Fahr. (775 degrees Cent.), all the specimens had approximately the same average grain size and the carbon was completely dissolved. With a cooling rate of 450 degrees Fahr. (250 degrees Cent.) per second, the austenite transformed in the Ar′ temperature range, whereas with a rate of 560 degrees Fahr. (310 degrees Cent.) per second, less than 10 per cent of the austenite decomposed at Ar′. With further progressive increase in cooling rate, the estimated proportion of the austenite that transformed in the Ar′ temperature range progressively decreased until none of the austenite transformed in this temperature range. Thus, there is a narrow range in quenching rates in which the austenite of high purity iron-carbon alloys transforms at Ar′ and a wide range in quenching rates in which the larger portion of the austenite transforms at Ar″. Critical cooling rate values, by definition, are restricted to the latter range in cooling rates. As shown by the slope of the transformation-cooling rate curve for the upper range of cooling rates, wide changes in cooling rates result in only minor changes in the proportion of troostite. For example, an increase in quenching rate from about 1000 to 1280 degrees Fahr. (555 to 710 degrees Cent.) per second decreased the estimated amounts of troostite only from 5 to 2 per cent. This fact should be borne in mind when the

critical cooling rate is used as the basis for a comparison of the hard-
enability of the alloys.

The hardness of the quenched specimens increased from Rock-
well C-40 to 65 as the cooling rate was increased from about 450 to
680 degrees Fahr. (250 to 375 degrees Cent.) per second. Further
increase in quenching rate had no marked effect on the hardness of
the alloy. Thus, appreciable proportions of troostite may be pro-
duced in the quenched specimens without its presence being indicated
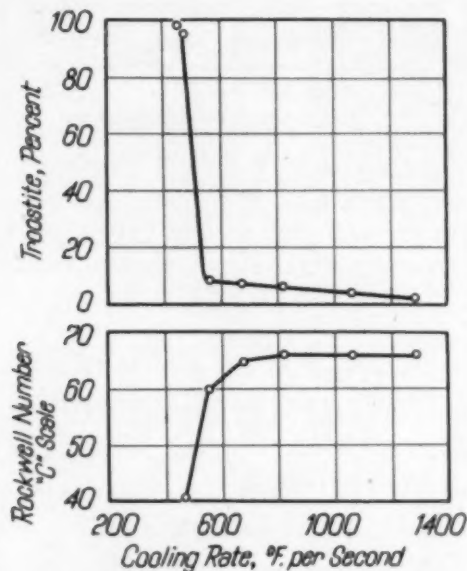
Fig. 11—Effect of Cooling Rate on
Structure and Hardness of Iron-Carbon
Alloy Containing 0.80 Per Cent Carbon
Quenched from 1425 Degrees Fahr. Ex-
cept for Cooling Rates Ranging from
450 to 560 Degrees Fahr. per Second
Where Corrections Were Made for the
Heat Effect Due to the Ar' Transforma-
tion, All Cooling Rates are Based on the
Average Rate Required for the Speci-
men to Cool Through the Temperature
Range 1110 to 930 Degrees Fahr. Hard-
ness Values Were Obtained on the Flat
Surface of the Quenched Specimens.

by hardness testing. This is shown by the difference in the slope of
the two curves in Fig. 11, for the upper range of quenching rates.
The Rockwell test indicated that the alloy was completely hardened
when the specimens contained troostite varying in amounts up to ap-
proximately 6 per cent. A similar relation was also shown in micro-
hardness testing. Obviously the hardness values alone, obtained on
the quenched specimens, could not be used as a precise index of the
actual cooling rate of the specimens.

The transition from the unhardened to the hardened condition

in high purity iron-carbon alloys is brought about by a small change
in cooling rate. This important fact is of particular significance in
considering the mechanism of the decomposition of austenite on
quenching the alloys.

The critical cooling rate was determined for the 0.80 per cent
carbon alloy quenched from temperatures ranging from 1425 to
1800 degrees Fahr. (775 to 980 degrees Cent.), for the 0.85 per cent
carbon alloy quenched from temperatures ranging from 1425 to
2000 degrees Fahr. (775 to 1095 degrees Cent.), and for the other
alloys ranging in carbon from 0.23 to 1.21 per cent quenched from
1800 degrees Fahr. (980 degrees Cent.).

The effect of the time to cool from 1110 to 930 degrees Fahr.
(600 to 500 degrees Cent.) on the structure of the 0.80 per cent
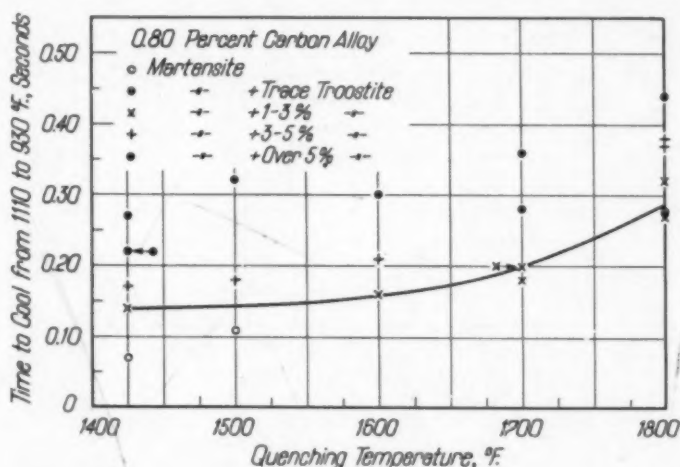
Fig. 12—Effect of Cooling Time on the Structure of the
0.80 Per Cent Carbon Alloy Quenched from Different Tem-
peratures. Specimens, 0.10 Inch Square by 0.040 Inch
Thick, Were Heated Rapidly in Nitrogen to the Temper-
atures of 1425, 1500, and 1600 Degrees Fahr., and in
Vacuo to 1800 Degrees Fahr., and Quenched in Hydrogen.
Specimens, 0.25 Inch Square by 0.040 Inch Thick, Were
Heated Rapidly in Nitrogen to the Temperature of 1700
Degrees Fahr. and Quenched Directly in a Sodium Silicate-
Water Bath.

carbon alloy quenched from various temperatures is shown in Fig.
12, and on the 0.85 per cent carbon alloy in Fig. 13. The effect of
time to cool from 1110 to 930 degrees Fahr. (600 to 500 Cent.) on
the structure of the alloys, varying in carbon content, quenched from
1800 degrees Fahr. (980 degrees Cent.) is shown in Fig. 14. The
relation of critical cooling time (resulting in 1 to 3 per cent troostite)
to quenching temperature is represented by the curves in Figs. 12
and 13, and to carbon content in Fig. 14.

The relation of critical cooling rate to quenching temperature
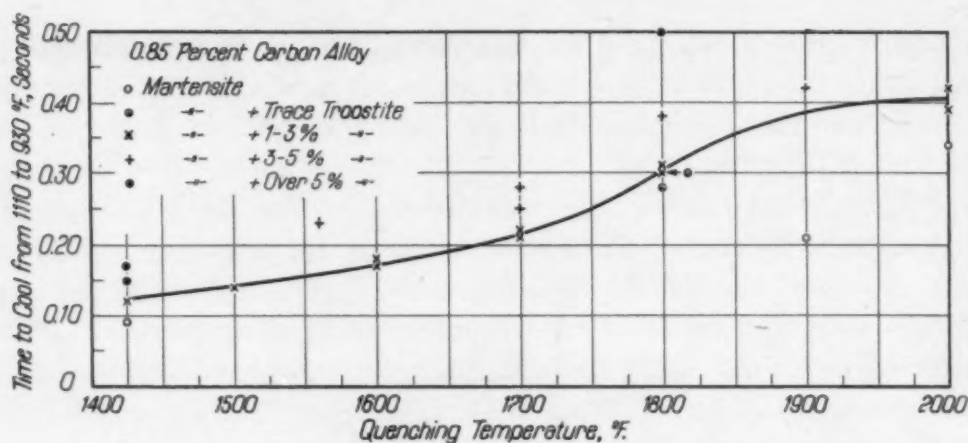
Fig. 13—Effect of Cooling Time on the Structure of the 0.85 Per Cent Carbon Alloy Quenched from Different Temperatures. Specimens, 0.10 Inch Square by 0.040 Inch Thick, were Heated Rapidly in Nitrogen to the Temperatures of 1425, 1500, and 1600 Degrees Fahr., and in Vacuo to 1800, 1900, and 2000 Degrees Fahr., and Quenched in Hydrogen. Specimens 0.25 Inch Square by 0.040 Inch Thick, Were Heated Rapidly in Nitrogen to 1700 Degrees Fahr. and Quenched in a Sodium Silicate-Water Bath.
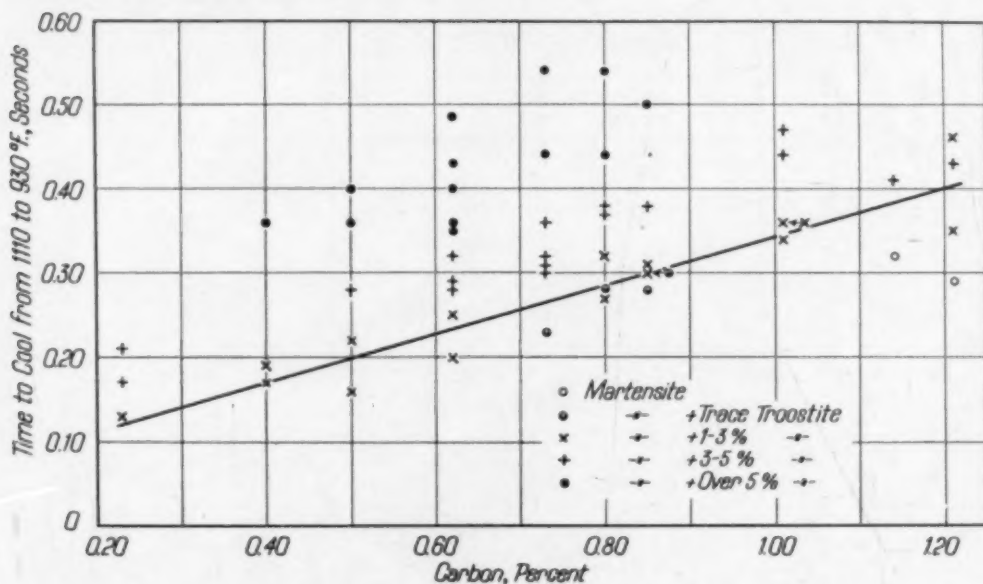
Fig. 14—Effect of Cooling Time on the Structure of Iron-Carbon Alloys. Specimens Were Heated Rapidly in Vacuo to 1800 Degrees Fahr. and Quenched Directly in Hydrogen. At 1800 Degrees Fahr., All the Alloys Had the Same Average Grain Size of 2.9 Grains Per Square Inch at 100 Diameters (Table II), and the Carbon Was Completely Dissolved.

for the alloys containing 0.80 to 0.85 per cent carbon is summarized in Fig. 15. These alloys were quenched from temperatures sufficiently high to insure complete solution of carbon and the time at these temperatures was sufficient for the carbon to be uniformly distributed. The decrease in critical cooling rate with increase in quenching temperature was due solely to variation in austenitic grain size.

INFLUENCE OF AUSTENITIC GRAIN SIZE AND CARBON CONTENT ON THE CRITICAL COOLING RATE

The data from Fig. 15 and Table II are assembled in Fig. 16 to show quantitatively the effect of variation in austenitic grain size on the critical cooling rate of the 0.80 to 0.85 per cent carbon alloys. The rate of heating through the transformation range to any selected temperature was of the same order of magnitude for specimens used in the determination of the critical cooling rate as that for specimens used in the measurement of the austenitic grain size. In some cases, the grain size was determined on the specimens used in the critical
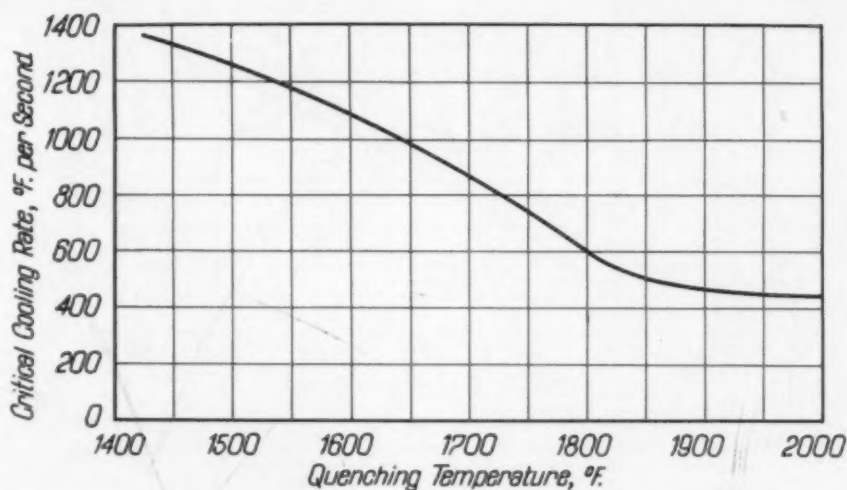
Fig. 15—Relation of Critical Cooling Rate to Quenching Temperature of Iron-Carbon Alloys Containing 0.80 to 0.85 Per Cent Carbon. The Austenitic Grain Size Ranged from Approximately 20 to 1½ Grains Per Square Inch at 100 Diameters.

cooling rate experiments. On quenching these alloys, the initial transformation of austenite to troostite occurred at the grain boundaries of the parent austenite. Some variations always existed in the size of the austenitic grains and only a small proportion of the austenite transformed to troostite on quenching at the critical cooling rate. It is obvious that precisely the same critical cooling rate values would not have been obtained had the alloys contained grains of only one size at each quenching temperature used. Correlation of the critical cooling rate with grain size is made on the basis of the average size of the austenite grains for each condition and not on the minimum or maximum values.

As shown in Fig. 16, the critical cooling rate progressively decreased (hardenability increased) with increase in austenitic grain size of the alloys containing 0.80 to 0.85 per cent carbon. For con-

stant carbon content and uniform distribution and solution of all carbon in austenite, the relation between critical cooling rate and austenitic grain size may be represented approximately by the empirical equation

$$R = \text{constant } (N)^n \tag{1}$$

where R = critical cooling rate in degrees Fahr. per second

N = number of austenitic grains per square inch at 100 diameters

n = constant = approximately 0.4
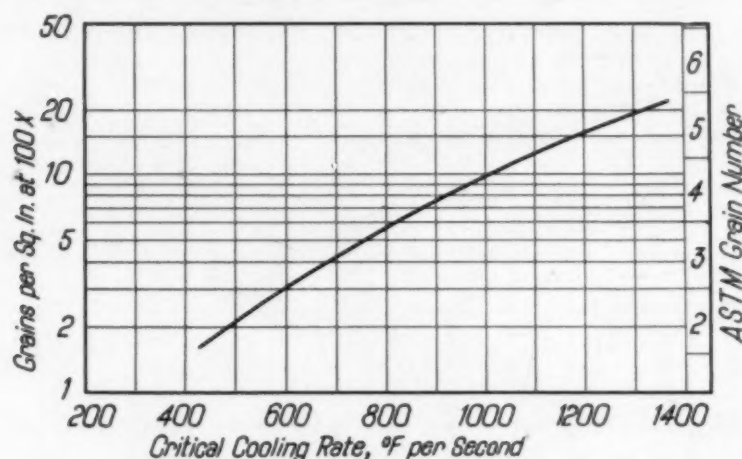
Fig. 16—Relation of Critical Cooling Rate to Austenitic Grain Size of Iron-Carbon Alloys Containing 0.80-0.85 Per Cent Carbon.

The basis for this equation is the fact that graphs of relationship between grain size and critical cooling rate, on logarithmic scale, are approximately straight lines.

The relation of critical cooling rate to carbon content of the alloys for two different austenitic grain sizes is shown in Fig. 17. The results summarized in curve A were obtained in previous experiments (1) by quenching specimens directly from 1700 degrees Fahr. (925 degrees Cent.), the temperature establishing an average constant grain size of approximately 7.5 grains per square inch at 100 diameters with all carbon in solution. The results summarized in curve B were obtained in the present experiments (Fig. 14) by quenching specimens directly from 1800 degrees Fahr. (980 degrees Cent.), the temperature establishing an average constant austenitic grain size of approximately 2.9 grains per square inch at 100 diameters with all carbon in solution. See Table II for grain size of the alloys at 1700 and 1800 degrees Fahr. (925 and 980 degrees Cent.).

The influence of austenitic grain size on the hardenability of the

alloys is also shown by a comparison of the results summarized in curves A and B. With any selected carbon content within the range of 0.23 to 1.21 per cent, the ratio of the critical cooling rate for an average austenitic grain size of 7.5 grains per square inch at 100 diameters to that for an average austenitic grain size of 2.9 grains per square inch at 100 diameters was approximately constant and
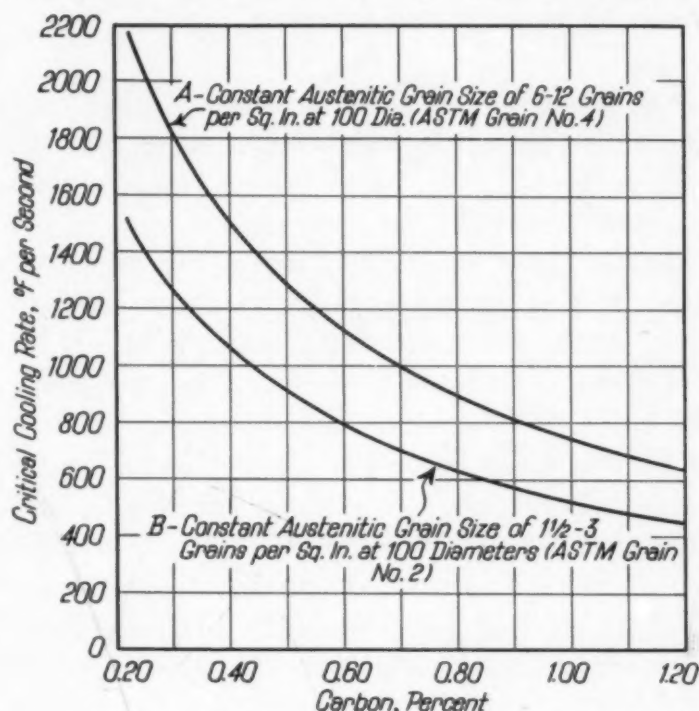
Fig. 17—Relation of Critical Cooling Rate to Carbon Content of Iron-Carbon Alloys.

was of the order of 1.43 to 1.0. That is, in order to attain complete hardening, an alloy with the smaller austenitic grains must be cooled approximately 1.5 times as fast as the same alloy with the larger grains.

For each austenitic grain size investigated (Fig. 17), the critical cooling rate decreased continuously with increase in carbon content of the alloys. For constant austenitic grain size and uniform distribution and solution of all carbon in austenite, the relation between critical cooling rate and carbon content of the alloys may be approximately represented by the empirical equation

$$R = \frac{constant}{C + c} \qquad (2)$$

where C = carbon content in per cent
      c = constant = 0.2.

The critical cooling rate, or hardenability, of the alloys depends only upon the austenitic grain size and carbon content, provided the quenching temperature is sufficiently high to dissolve the carbon completely and the time at the quenching temperature is sufficient to insure its uniform distribution. The hardenability of the alloys, therefore, may be expressed in terms of an empirical equation by combining equations (1) and (2) as follows:

$$R = \frac{K\,(N)^n}{C + c} \qquad\qquad (3)$$

$$R = \frac{410\,N^{0.4}}{C + 0.2} \qquad\qquad (4)$$

where K = constant = 410.

This empirical equation is based upon experiments with high purity iron-carbon alloys ranging in carbon from 0.23 to 1.21 per cent and in austenitic grain size from about 20 to 1½ grains per square inch at 100 diameters. It, therefore, has not been shown to be applicable for computing the critical cooling rate of ordinary steels, or for computations involving alloys of very low carbon content. For example, equation (4) indicates that high purity iron can be made martensitic by fast quenching, obviously a condition that is impossible. It should be pointed out that the determinations for carbon were made with considerably more precision than was possible in the determinations of the austenitic grain size of the alloys.

Surface Hardness

French and Klopsch (4) reported the results of Rockwell hardness tests made on specimens of carbon steels varying in carbon content and quenched at different rates. For specimens quenched at the critical cooling rate, hardness increased rapidly with carbon content up to about 0.70 per cent but thereafter remained constant with increase in carbon. With steels containing 0.75 to 1.25 per cent carbon, practically identical hardness values were obtained on the quenched specimens containing appreciable proportions of troostite and on the completely martensitic specimens. In 1938, Burns, Moore, and Archer (11) showed that the maximum hardness attainable by quenching the plain carbon and common alloy steels depends alone on the carbon content. The hardness values (Rockwell C) increased with carbon up to about 0.55 per cent and thereafter remained constant with increase in carbon (for the range in carbon investigated).

In the present investigation, readings for hardness were made with the Rockwell Superficial tester (30-kilogram load) on most of the specimens used in the determination of the critical cooling rate of the alloys. For comparison purposes, the readings were converted to the C scale in accordance with the conversion tables supplied by the manufacturer of the instrument. The flat surfaces of the quenched specimens were cleaned with 1-G emery paper before testing.

The results of the hardness tests made on specimens of the alloys quenched directly from 1800 degrees Fahr. (980 degrees Cent.) at various rates are summarized in Fig. 18. The estimated amounts of troostite formed (1 to 3 per cent for critical cooling rate) are also
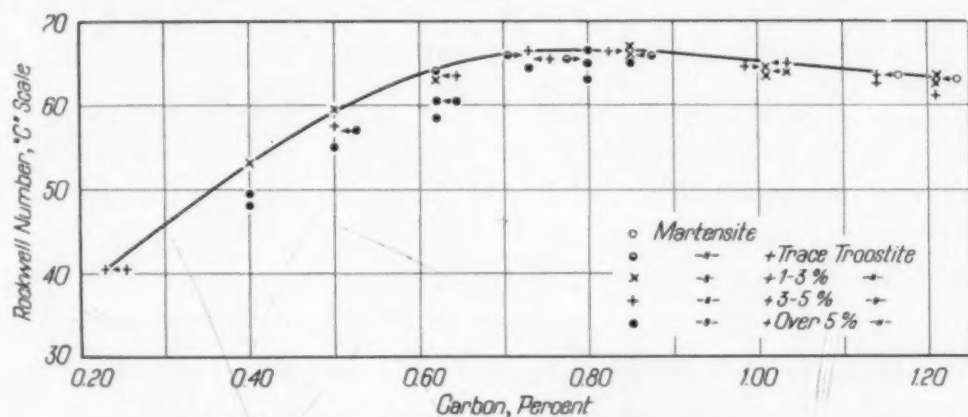
Fig. 18—Effect of Carbon Content on the Hardness of Iron-Carbon Alloys.

shown, together with the curve representing the relation of maximum values obtained for hardness to carbon content.

The alloys had the same average grain size of 2.9 grains per square inch at 100 diameters and all carbon was in solution at the time of quenching, but all specimens were not completely martensitic after the quench as evidenced by the presence of troostite. The latter specimens, with troostite estimated up to 5 per cent, had Rockwell hardness values of the same order of magnitude as the fully martensitic specimens of similar carbon content. The maximum hardness values obtained increased with increase in carbon up to about 0.70 per cent, remained constant for alloys ranging in carbon from 0.70 to 0.85 per cent, and decreased slightly with further increase in carbon. The decrease in hardness with increase in carbon of the alloys in the high carbon range is believed to be due to the retention of some austenite after the quench. The maximum hardness values

obtained for the alloys in the low carbon range are considerably lower than those reported by Burns, Moore, and Archer (11) for steels of similar carbon content, whereas in the high carbon range the alloys had approximately the same hardness values as the steels.

As is shown in Fig. 19, the maximum hardness value for the 0.85 per cent carbon alloy was not affected by variation in quenching temperatures from 1425 to 2000 degrees Fahr. (775 to 1095 degrees Cent.). This range in quenching temperatures produced austenitic grain sizes ranging from approximately 20 to 1½ grains per square inch at 100 diameters (Table II). The maximum hardness value obtained for this alloy, therefore, was not influenced by wide variations in the grain size at the time of quenching.

SUMMARY

The apparatus and procedure for heating small specimens in vacuo and in an atmosphere of dry nitrogen to various temperatures followed by quenching directly in hydrogen are described. Oxidation of the specimen during the time of heating and quenching was eliminated, and it was possible to obtain progressive changes in cooling
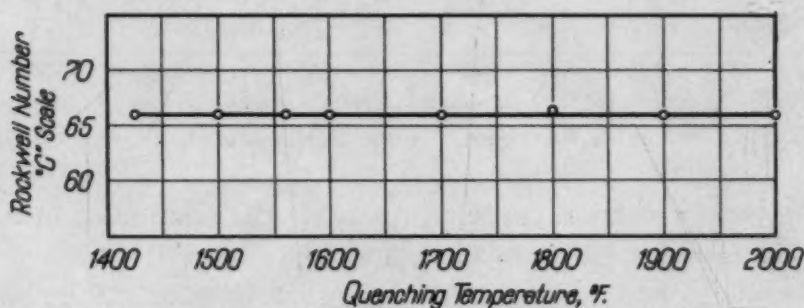
Fig. 19—Effect of Quenching Temperature on the Hardness of the Iron-Carbon Alloy Containing 0.85 Per Cent Carbon.

rates over the wide range required in the present experiments. Specimens quenched in hydrogen were cooled uniformly over the entire surface, as shown by the microstructure in the cross section.

The string galvanometer apparatus and the procedure used for obtaining photographic records of the time-temperature cooling curves are also described in some detail. Examples are given of typical cooling curves obtained on specimens heated in vacuo and in an atmosphere of dry nitrogen and quenched directly in hydrogen.

The high purity iron-carbon alloys used in the present study ranged in carbon from 0.23 to 1.21 per cent. The specimens were

from the same alloys as those used and described in previous investigations (1), (2).

The grain size established at temperatures ranging from 1425 to 1600 degrees Fahr. (775 to 870 degrees Cent.) increased markedly with decrease in rate of heating through the transformation range, whereas the grain size at 1800 degrees Fahr. (980 degrees Cent.) was not so noticeably dependent upon the rate of heating. The grain size at temperature ranging from 1425 to 2100 degrees Fahr. (775 to 1150 degrees Cent.) was determined for the alloys. For any selected temperatures within the range from 1600 to 2100 degrees Fahr. (870 to 1150 degrees Cent.) and with the heating rate employed, all the alloys were found to have approximately the same average grain size provided the carbon was completely dissolved at that temperature. Carbon in solution, therefore, was not effective in inhibiting grain growth of the austenite in the high purity iron-carbon alloys. The actual temperatures and the rate of heating through the transformation range were the dominant factors in controlling the austenitic grain size of these alloys.

There is a narrow range in cooling rates in which the austenite transforms at Ar′ and a wide range in quenching rates in which the larger portion of the austenite transforms at Ar″. Thus, the transition from the unhardened to the hardened condition in high purity iron-carbon alloys is brought about by a small change in cooling rate.

The critical cooling rate was determined for the 0.80 per cent carbon alloy quenched directly from temperatures ranging from 1425 to 1800 degrees Fahr. (775 to 980 degrees Cent.), for the 0.85 per cent carbon alloy quenched from temperatures ranging from 1425 to 2000 degrees Fahr. (775 to 1095 degrees Cent.), and for the alloys varying in carbon from 0.23 to 1.21 per cent quenched from 1800 degrees Fahr. (980 degrees Cent.).

The critical cooling rate was taken as the average cooling rate, between 1110 and 930 degrees Fahr. (600 and 500 degrees Cent.), which produced in the quenched specimen a structure of martensite with nodular troostite (fine pearlite) in amounts estimated to be between 1 and 3 per cent.

The critical cooling rate progressively decreased (hardenability increased) as the austenitic grain size increased in the alloys containing 0.80 to 0.85 per cent carbon. For each austenitic grain size investigated, the critical cooling rate decreased continuously with increase in carbon content of the alloys.

For complete solution and uniform distribution of carbon at the time of quenching, the critical cooling rate (R in degrees Fahr. per second) or hardenability of the alloys may be approximately represented by the equation

$$R = \frac{410\ N^{0.4}}{C + 0.2}$$

Where N = number of austenitic grains per square inch at 100 diameters

and C = carbon content in per cent.

Hardness values obtained on the quenched specimens could not alone be used as a precise index to the actual cooling rates of the alloys.

The maximum hardness values obtained on specimens of each alloy quenched from 1800 degrees Fahr. (980 degrees Cent.) increased with increase in carbon up to about 0.70 per cent, remained constant for alloys ranging from 0.70 to 0.85 per cent, and decreased slightly with further increase in carbon. The maximum hardness value for the 0.85 per cent carbon alloy was not affected by variation in quenching temperatures from 1425 to 2000 degrees Fahr. (775 to 1095 degrees Cent.). This range in quenching temperatures produced in the alloy austenitic grain sizes ranging from approximately 20 to 1½ grains per square inch at 100 diameters.

Variation in observed cooling rates from about 550 to 2600 degrees Fahr. (305 to 1445 degrees Cent.) per second with austenitic grain size of approximately 20 grains per square inch at 100 diameters had no appreciable influence on the temperature range of the start of the Ar" transformation in the 0.80 per cent carbon alloy. Variation in austenitic grain size ranging from approximately 20 to 3 grains per square inch at 100 diameters had no detectable effect on the temperature of the start of the Ar" in this alloy.

Grateful acknowledgment is due to N. L. Carwile and S. J. Rosenberg for assistance in this work.

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Oral Discussion

ROBERT M. PARKE:[1] I regret that I have not examined this interesting paper by Mr. Digges with the attention it warrants. One question occurs to me, although it may not be particularly significant. Has Mr. Digges given consideration to the possibility of the effect of absorption of hydrogen by the austenitized steel during quenching on the rate of austenite transformation?

B. R. QUENEAU:[2] One point that I would like to bring out is that Fig. 16, showing the effect of austenite grain size on the critical cooling rate, does not follow directly from a knowledge of the effect of quenching temperatures, Fig. 15. The assumption that samples heated to different temperatures are alike in everything but grain size is probably incorrect. Recent studies have indicated that so-called "germ nuclei" exist in the austenite at temperatures above the critical which act as nuclei for subsequent transformation on cooling. The effective number of these "germ nuclei" decrease with increase in temperature and holding time and the higher the austenitizing temperature, the slower the rate of decomposition.

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[1] Research staff, Climax Molybdenum Co., Detroit.
[2] Assistant professor of metallurgy, Columbia University, New York.

It is well to bear in mind that both grain size and prior heat treatment affect the critical cooling rate. We would expect that the effect of grain size on critical cooling rate shown in Fig. 16 would be modified if these two factors had been separated.

Authors' Reply

Possibly some hydrogen could have been absorbed by the specimens, especially in the surface layers, during the time of quenching in hydrogen. However, the time the austenite was exposed to the hydrogen was relatively short due to the very rapid cooling rates required to prevent the austenite of these high-purity alloys from transforming to troostite. The structures of the specimens quenched in hydrogen also showed uniformity in the transformation products throughout the cross sections. Comparisons of the results obtained on specimens quenched in sodium silicate-water bath with specimens quenched in hydrogen indicated that the absorption of hydrogen had no detectable effect on the reported values of the critical cooling rate.

Since all specimens were treated alike, the results of the experiments summarized in this report were not influenced by variations in prior working or heat treatments. The critical cooling rate was determined for the alloys containing 0.80 and 0.85 per cent carbon quenched directly from 1700 degrees Fahr. The critical cooling rate was also determined for the same alloys under similar conditions except that after establishing the grain size at 1700 degrees Fahr. the specimens were cooled to 1400 degrees Fahr. and held for 5 minutes before quenching. Although these specimens were quenched from different temperatures (1400 and 1700 degrees Fahr.) the average grain size in each specimen was the same, namely, that established at 1700 degrees Fahr. The critical cooling rate of each alloy quenched directly from 1700 degrees Fahr. was the same as that for the corresponding alloy cooled from 1700 to 1400 degrees Fahr. before quenching. This would indicate that the quenching temperature was important only in establishing the austenitic grain size provided the temperature is sufficiently high to dissolve all the carbon and time at temperature is sufficient to insure its uniform distribution. It would seem, therefore, that the results summarized in Fig. 16 were not influenced by variations in so-called "germ nuclei".

THE EFFECT OF MOLYBDENUM ON THE ISOTHERMAL, SUBCRITICAL TRANSFORMATION OF AUSTENITE IN LOW AND MEDIUM CARBON STEELS

By James R. Blanchard, Robert M. Parke, and Alvin J. Herzig

Abstract

From observations of the constant temperature, subcritical transformation of austenite, S-curves were constructed for steels in low and medium carbon ranges and with molybdenum contents from 0.15 to 0.75 per cent. The effect of molybdenum on the microconstituents produced at various isothermal transformation temperatures is discussed.

Introduction

THE original work on the isothermal transformation of steel at subcritical temperatures by Davenport and Bain (1),[1] published ten years ago, provided a theory and a technique in ferrous metallurgy which have been of exceptional value to the metallurgist in the development and control of heat treating processes for iron-base alloys. Since the isothermal technique also provides a means for the study of the fundamental effects of alloys on the hardening reactions in steel, an investigation was begun to determine the effect of molybdenum on the isothermal transformation of austenite at subcritical temperatures. This report presents the data obtained for steels of low and medium carbon content.

Previous investigators have found that, of the many factors which influence the rate of austenite decomposition at constant subcritical temperatures, the most noteworthy are: (a) austenite grain size, (b) rate of nucleation, (c) rate of growth, (d) undissolved carbides and inclusions, (e) degree of heterogeneity, (f) chemical composition.

These factors are somewhat interdependent, but they have been

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[1]The figures appearing in parentheses refer to the bibliography appended to this paper.

discussed in detail by Davenport (2)[2] and Mehl (3); so it suffices
here to review the fact that, of the six, chemical composition of the
austenite exerts the most important effect on the time for initiation
and rate of transformation in steel.

PREPARATION OF STEELS AND PROCEDURE

This investigation is concerned with the effects of 0.15-0.75
per cent molybdenum on the isothermal, subcritical transformation of
steel in two carbon ranges: (a) 0.11-0.12 per cent carbon, (b)
0.39-0.45 per cent carbon. Accordingly, two series of steels were
made in a high frequency induction furnace. The analyses are given
in Table I.

Table I

No.	C Per Cent	Mo Per Cent	Mn Per Cent	Si Per Cent	S Per Cent	P Per Cent	Ni Per Cent
1	0.12	0.008	0.65	0.26	0.012	0.009	0.06
2	0.12	0.13	0.61	0.27	0.012	0.009	0.06
3	0.12	0.31	0.63	0.25	0.012	0.009	0.06
4	0.12	0.49	0.60	0.22	0.012	0.009	0.06
5	0.11	0.74	0.59	0.25	0.012	0.009	0.06
6	0.45	0.009	0.67	0.26	0.012	0.009	0.06
7	0.44	0.15	0.65	0.28	0.012	0.009	0.06
8	0.41	0.29	0.69	0.23	0.012	0.009	0.06
9	0.39	0.49	0.67	0.26	0.012	0.009	0.06
10	0.42	0.71	0.55	0.25	0.012	0.009	0.06

The steels were melted in an acid-lined furnace, aluminum-killed,
and cast into fifteen-pound ingots. The ingots were forged into one-
inch square bars which were rolled into one-half inch wide strips
0.070-inch thick. The strips were normalized at 1700 degrees Fahr.
(870 degrees Cent.) and annealed at 1500 degrees Fahr. (815 de-
grees Cent.).

The transformations were followed metallographically at tem-
peratures above 600 degrees Fahr. (315 degrees Cent.) and
dilatometrically at temperatures below. The specimens used for the
metallographic method were ½ inch x ½ inch x 0.050 inch, and the
dilatometer specimens were 4⅜ inches long, ½ inch wide and 0.035
inch thick.

The steels in the 0.42 per cent carbon group were heated at a
solution temperature of 1550 degrees Fahr. (845 degrees Cent.) for
one-half hour, and the steels in the 0.12 per cent carbon group were
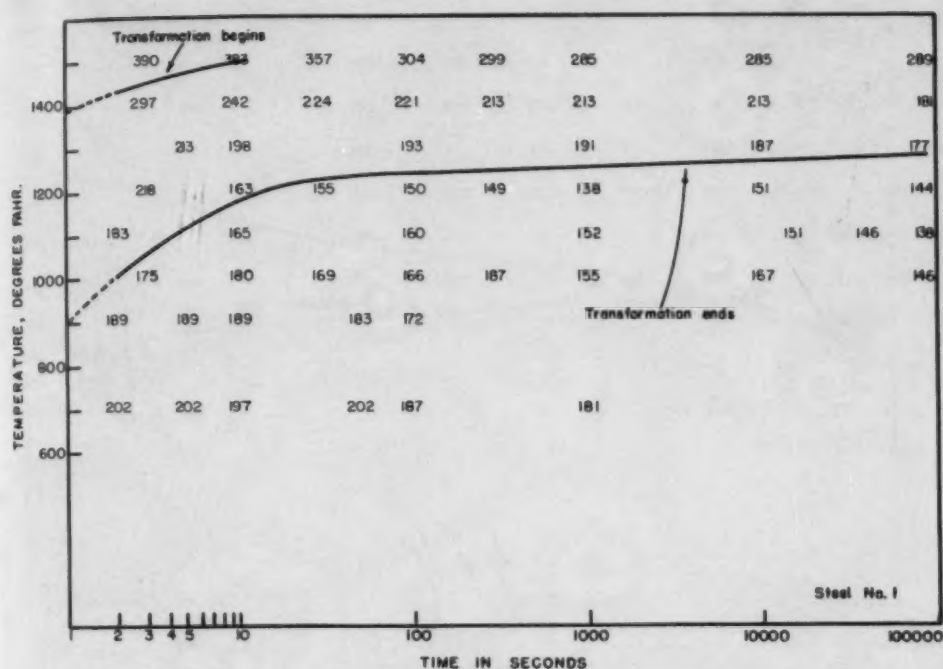
Fig. 1—Steel No. 1.   Carbon 0.12 Per Cent, Molybdenum 0.008 Per Cent.
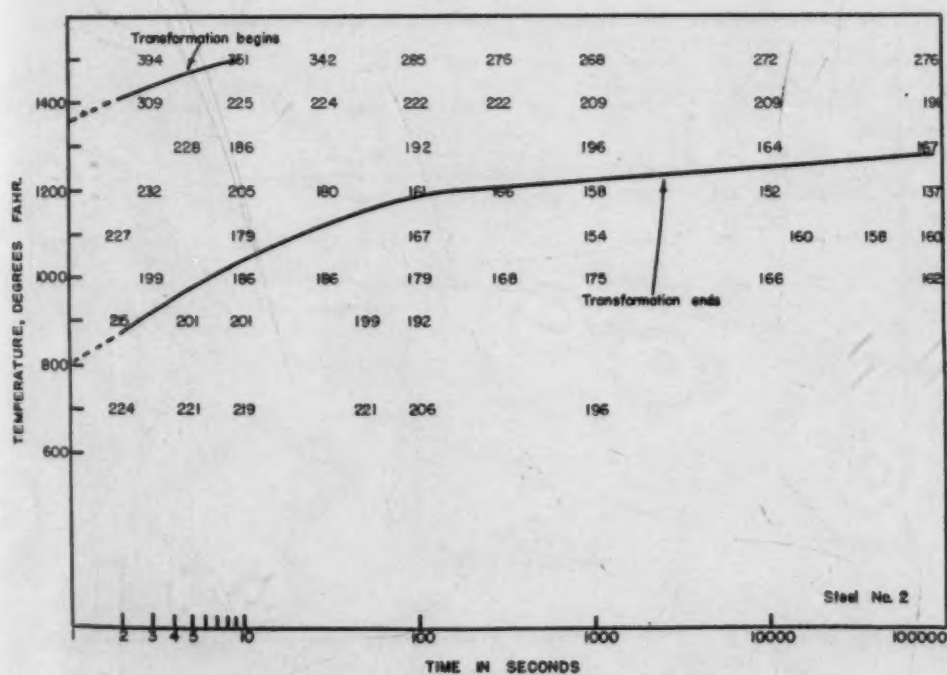
Fig. 2—Steel No. 2.   Carbon 0.12 Per Cent, Molybdenum 0.13 Per Cent.
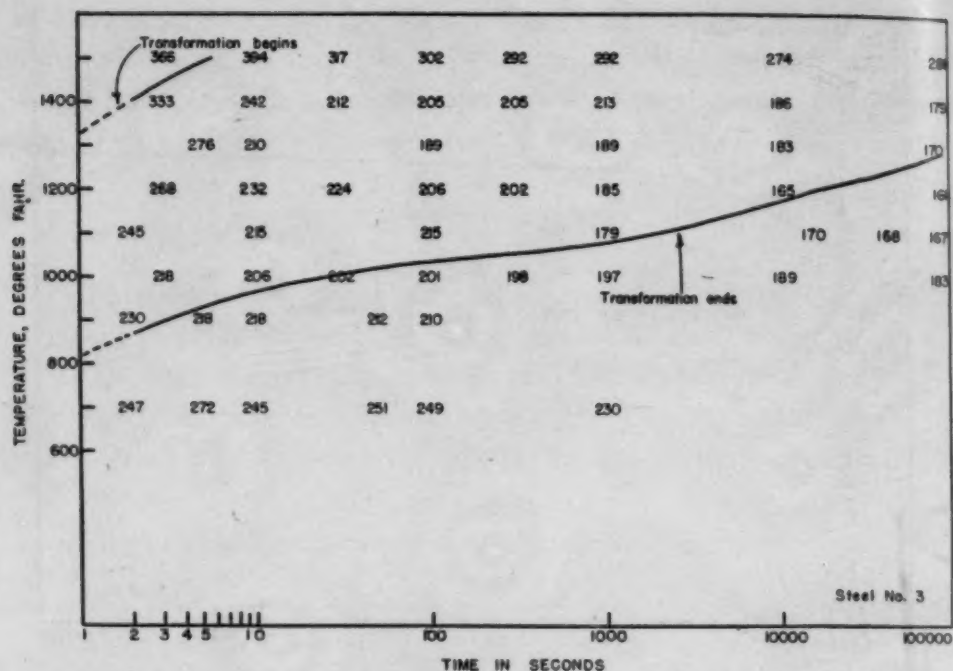
Fig. 3—Steel No. 3.　Carbon 0.12 Per Cent, Molybdenum 0.31 Per Cent.
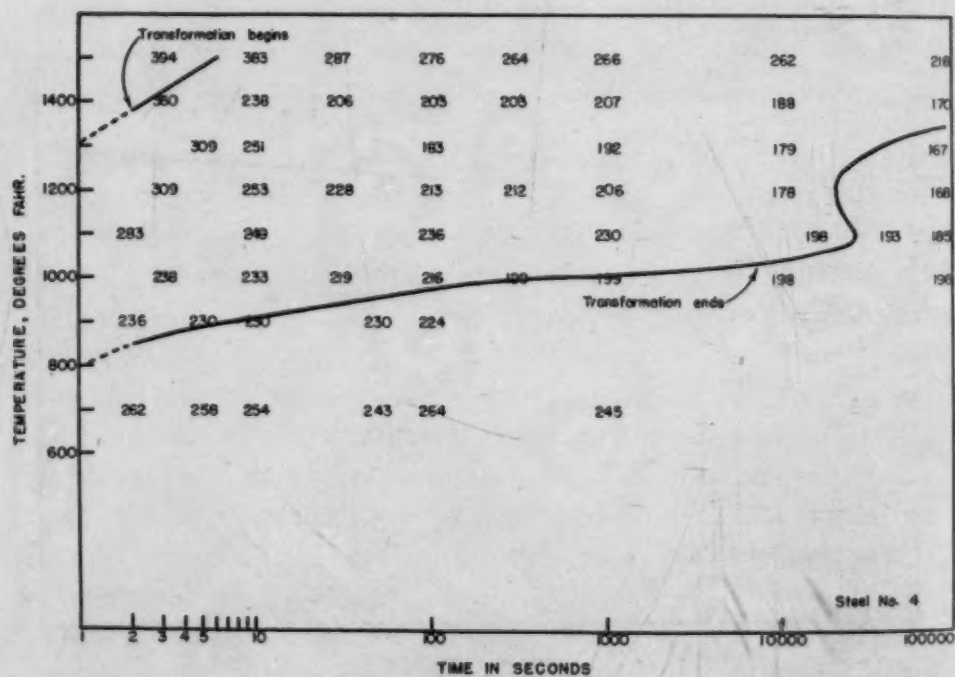
Fig. 4—Steel No. 4.　Carbon 0.12 Per Cent, Molybdenum 0.49 Per Cent.
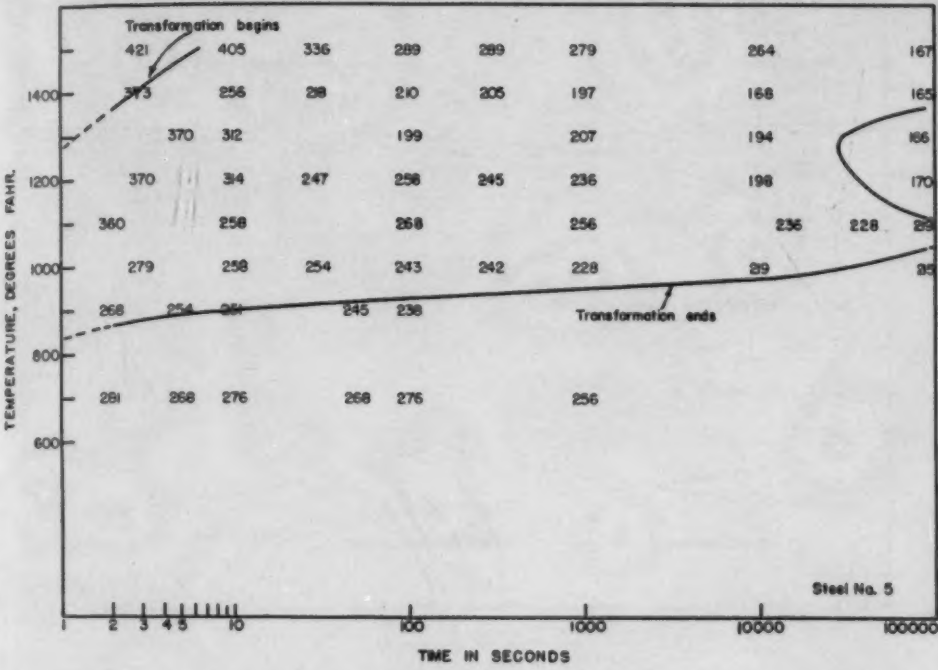
Fig. 5—Steel No. 5. Carbon 0.11 Per Cent, Molybdenum 0.74 Per Cent.
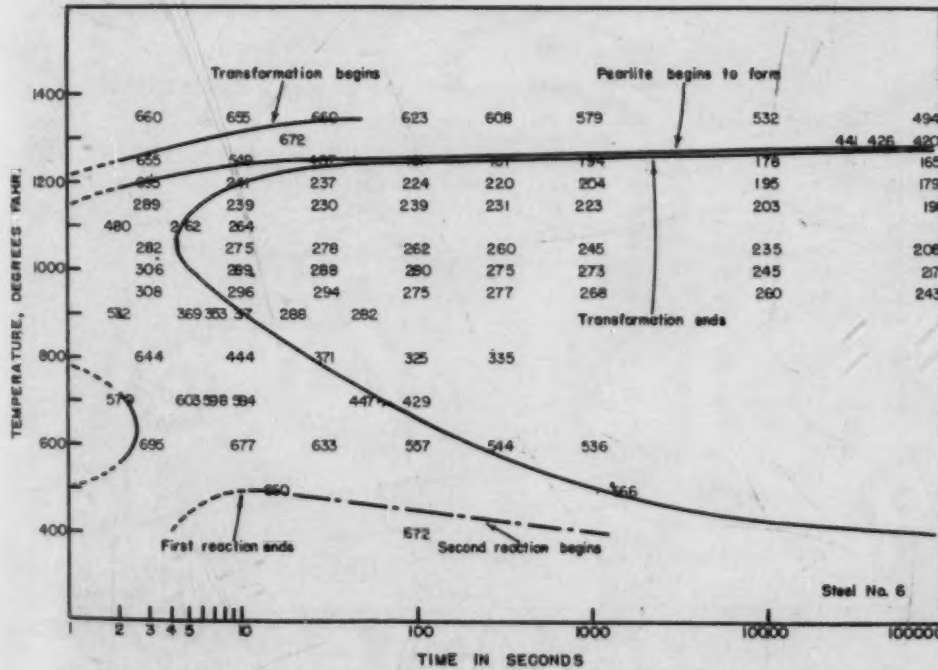
Fig. 6—Steel No. 6. Carbon 0.45 Per Cent, Molybdenum 0.009 Per Cent.
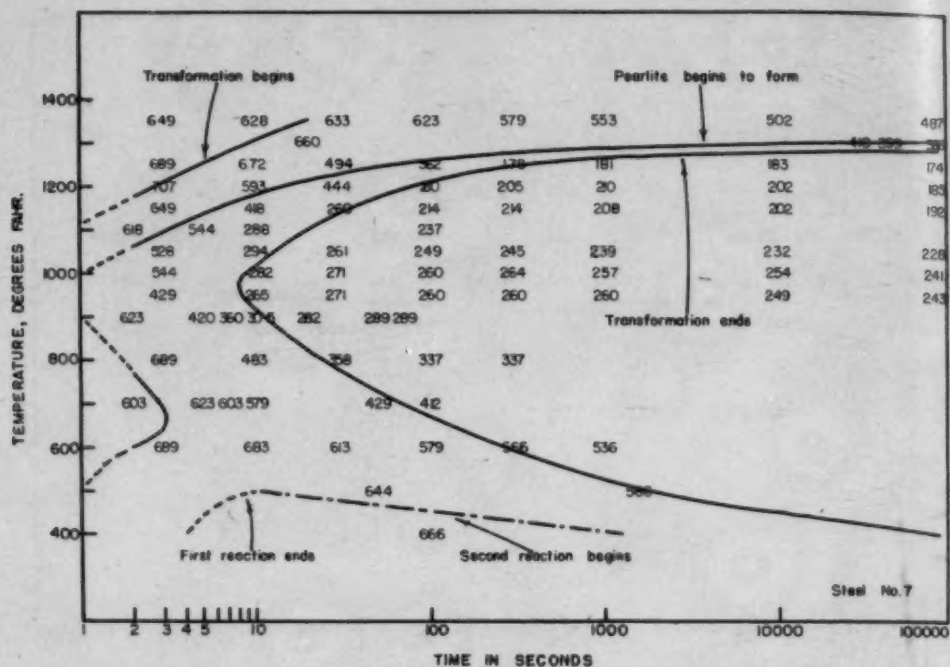
Fig. 7—Steel No. 7.   Carbon 0.44 Per Cent, Molybdenum 0.15 Per Cent.
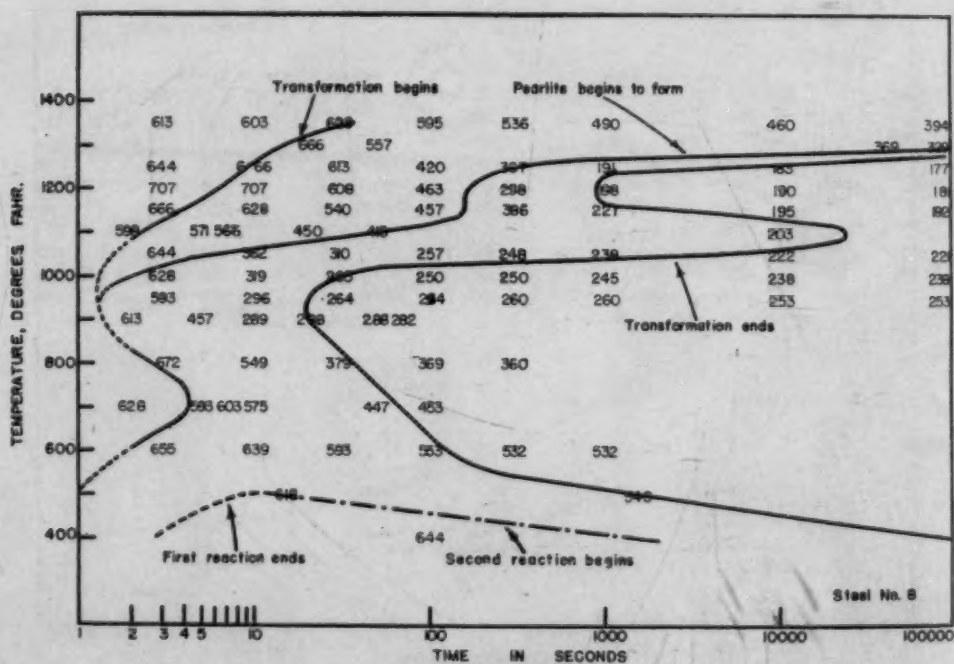
Fig. 8—Steel No. 8.   Carbon 0.41 Per Cent, Molybdenum 0.29 Per Cent.
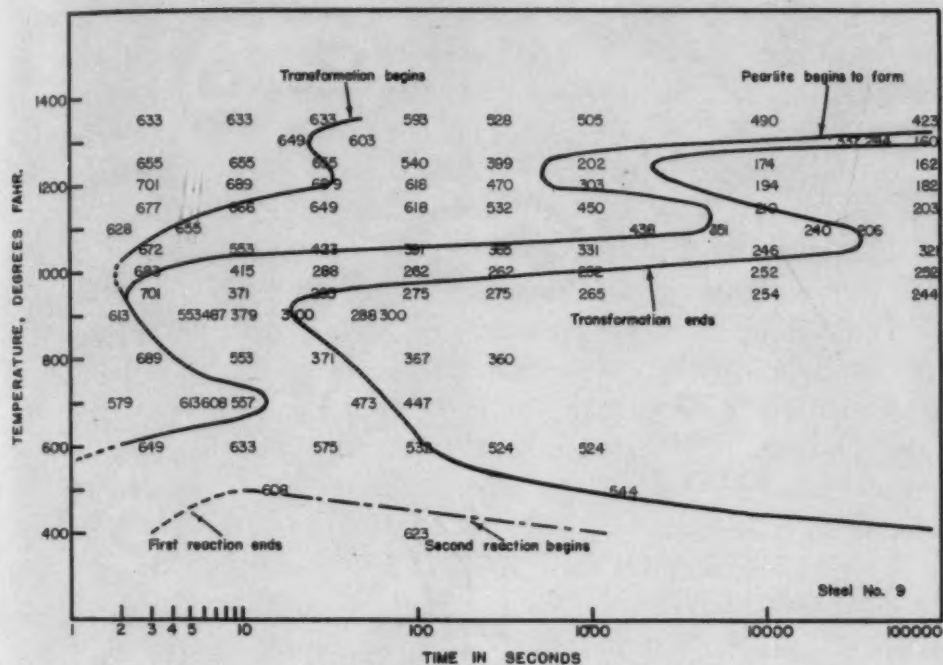
Fig. 9—Steel No. 9.   Carbon 0.39 Per Cent, Molybdenum 0.49 Per Cent.
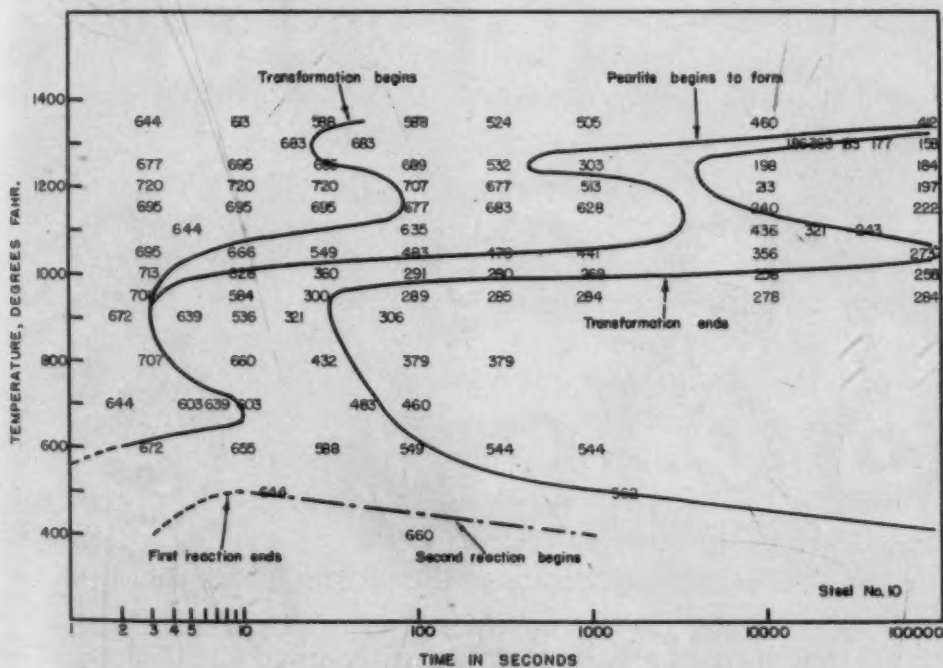
Fig. 10—Steel No. 10.   Carbon 0.42 Per Cent, Molybdenum 0.71 Per Cent.

heated at a solution temperature of 1700 degrees Fahr. (925 degrees Cent.) for one-half hour. The isothermal treatment was conducted in molten metal baths at intervals in the range 400 to 1350 degrees Fahr. (205 to 730 degrees Cent.) for the 0.42 per cent carbon series and in the range 700 to 1500 degrees Fahr. (370 to 815 degrees Cent.) for the 0.12 per cent carbon series.

The quenching dilatometer used with the 0.42 per cent carbon steels has been described elsewhere (4).

The data obtained are presented in Figs. 1 to 10 inclusive in the form of S-curves. By convention the curve for the initiation of transformation represents the time required for the first 1 per cent of the austenite to transform. Likewise, the curve marking the end of transformation denotes the time required for 99 per cent of the austenite to transform.

The broken lines between 400 and 500 degrees Fahr. (205 and 260 degrees Cent.) in the curves for the medium carbon steels, marked "End of First Reaction" and "Beginning of Second Reaction", designate a period of comparative inactivity in the breakdown of austenite to an acicular structure. It must not be assumed that these terms indicate that two separate reactions exist in this region. They are used to call attention to a definite change in the rate of austenite transformation.

The S-curve form for the presentation of isothermal data was selected because the method is well established in this country and appears to be more readily interpreted than others previously used.

The numbers on the S-curve charts are the room temperature Vickers diamond pyramid hardness numbers of the specimens which had been held for the designated time interval and at the temperature corresponding to the position of the numbers on the charts. It must be borne in mind that the hardness values of the incompletely transformed specimens are to a certain extent a reflection of the amounts of martensite in the specimen. The hardness numbers serve as a rough verification of the start and finish of transformation and are useful as an indication of the hardness one can expect when these ten steels are isothermally treated at the designated temperatures and times.

The microstructures for the plain carbon and the 0.75 per cent molybdenum steels are presented in Figs. 11 to 36. These structures show the influence of molybdenum on the constituents formed at various transformation temperatures.

Fig. 11—Carbon 0.45 Per Cent, Molybdenum 0.009 Per Cent. Held 100 Seconds at 1250 Degrees Fahr. Completely Transformed. × 1000.

Fig. 12—Carbon 0.42 Per Cent, Molybdenum 0.71 Per Cent. Held 10,000 Seconds at 1250 Degrees Fahr. Completely Transformed. × 1000.

Fig. 13—Carbon 0.45 Per Cent, Molybdenum 0.009 Per Cent. Held 10 Seconds at 1150 Degrees Fahr. Completely Transformed. × 1000.

Fig. 14—Carbon 0.42 Per Cent, Molybdenum 0.71 Per Cent. Held 10,000 Seconds at 1150 Degrees Fahr. Completely Transformed. × 1000.
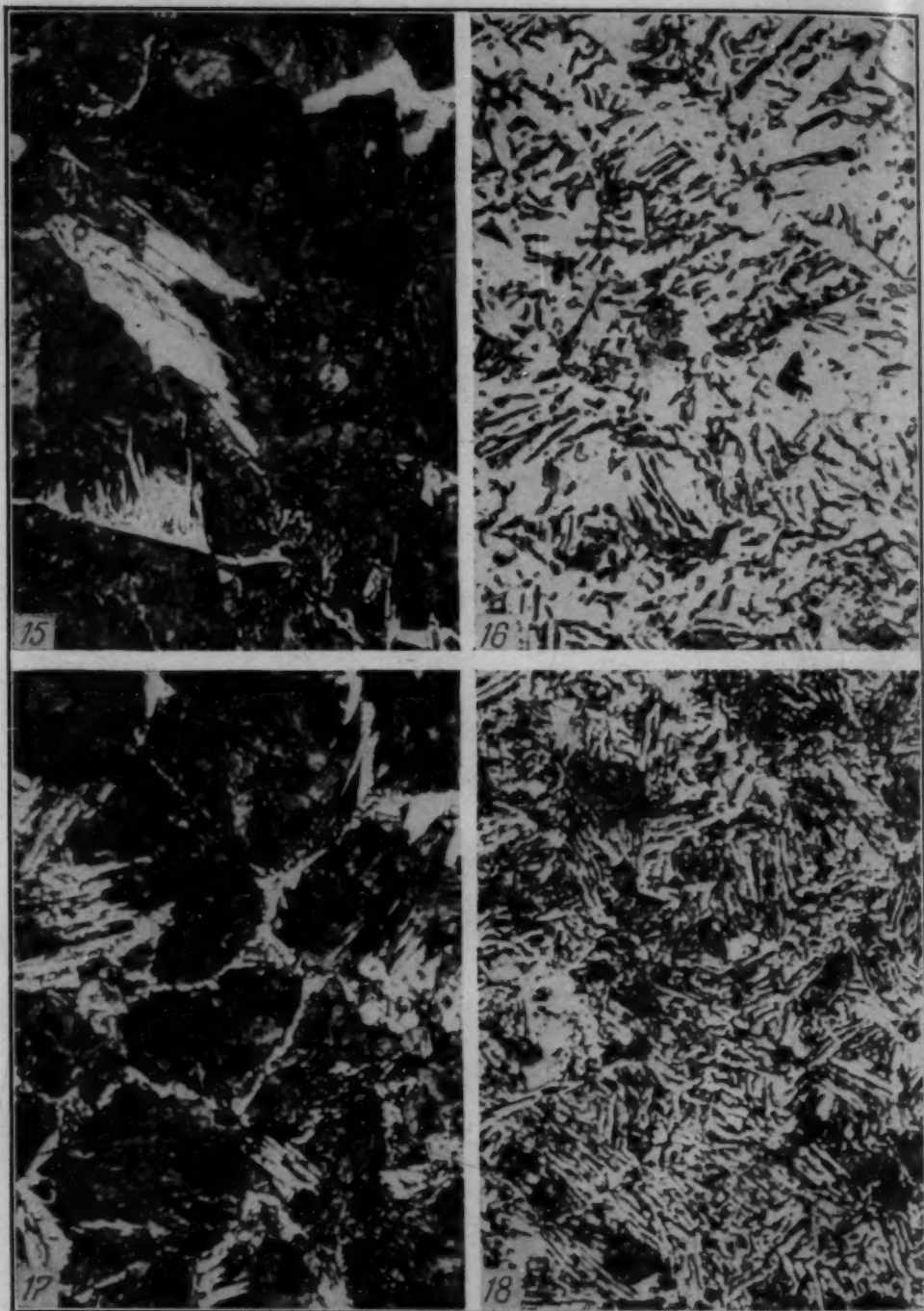
Fig. 15—Carbon 0.45 Per Cent, Molybdenum 0.009 Per Cent. Held 3 Seconds at 1000 Degrees Fahr. 97.7 Per Cent Transformed. × 1000.
Fig. 16—Carbon 0.42 Per Cent, Molybdenum 0.71 Per Cent. Held 10,000 Seconds at 1000 Degrees Fahr. Completely Transformed. × 1000.
Fig. 17—Carbon 0.45 Per Cent, Molybdenum 0.009 Per Cent. Held 10 Seconds at 950 Degrees Fahr. Completely Transformed. × 1000.
Fig. 18—Carbon 0.42 Per Cent, Molybdenum 0.71 Per Cent. Held 30 Seconds at 950 Degrees Fahr. Completely Transformed. × 1000.
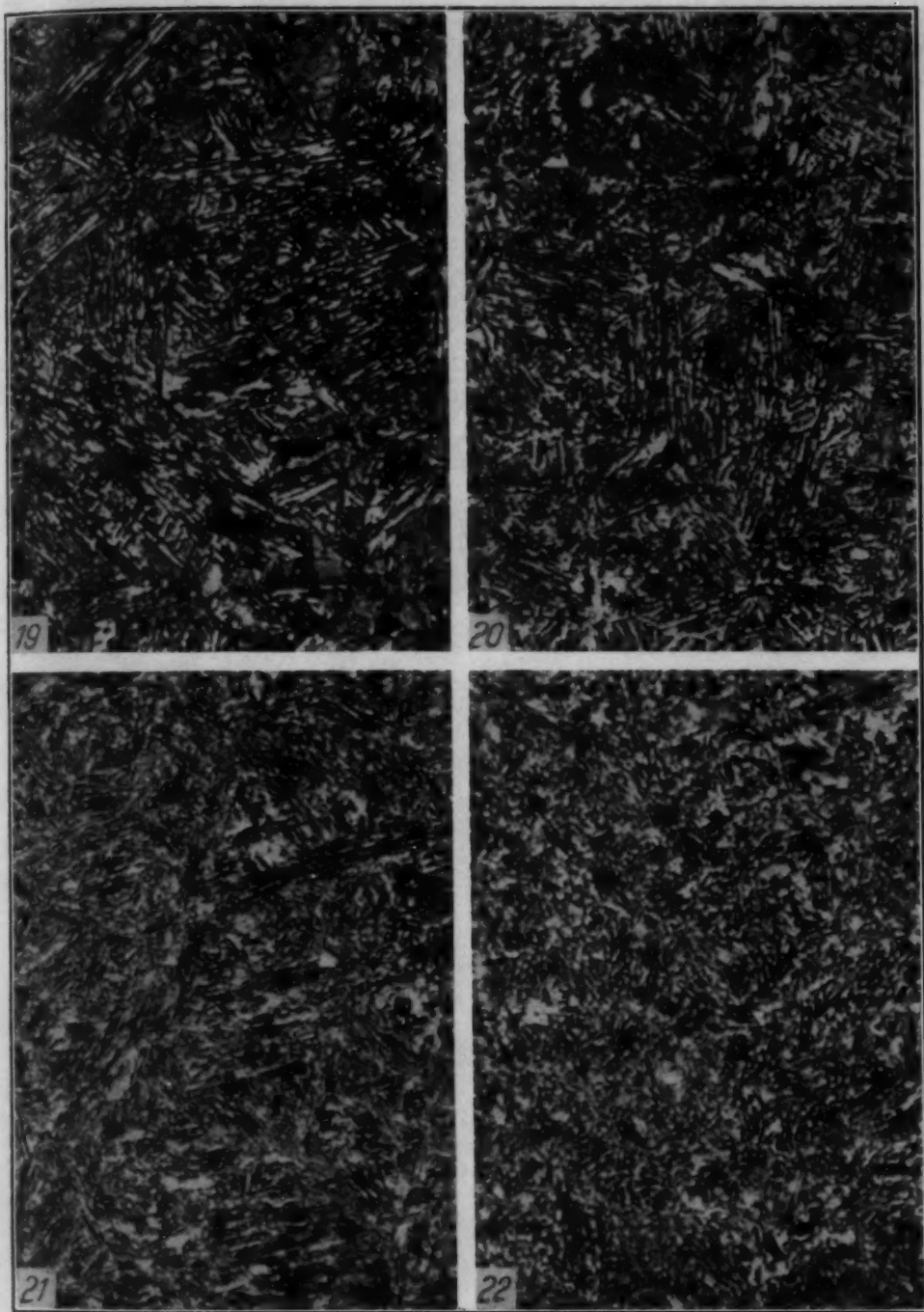
Fig. 19—Carbon 0.45 Per Cent, Molybdenum 0.009 Per Cent. Held 20 Seconds at 900 Degrees Fahr. Completely Transformed. × 1000.
Fig. 20—Carbon 0.42 Per Cent, Molybdenum 0.71 Per Cent. Held 50 Seconds at 900 Degrees Fahr. Completely Transformed. × 1000.
Fig. 21—Carbon 0.45 Per Cent, Molybdenum 0.009 Per Cent. Held 1500 Seconds at 500 Degrees Fahr. Completely Transformed. × 1000.
Fig. 22—Carbon 0.42 Per Cent, Molybdenum 0.71 Per Cent. Held 1500 Seconds at 500 Degrees Fahr. Completely Transformed. × 1000.
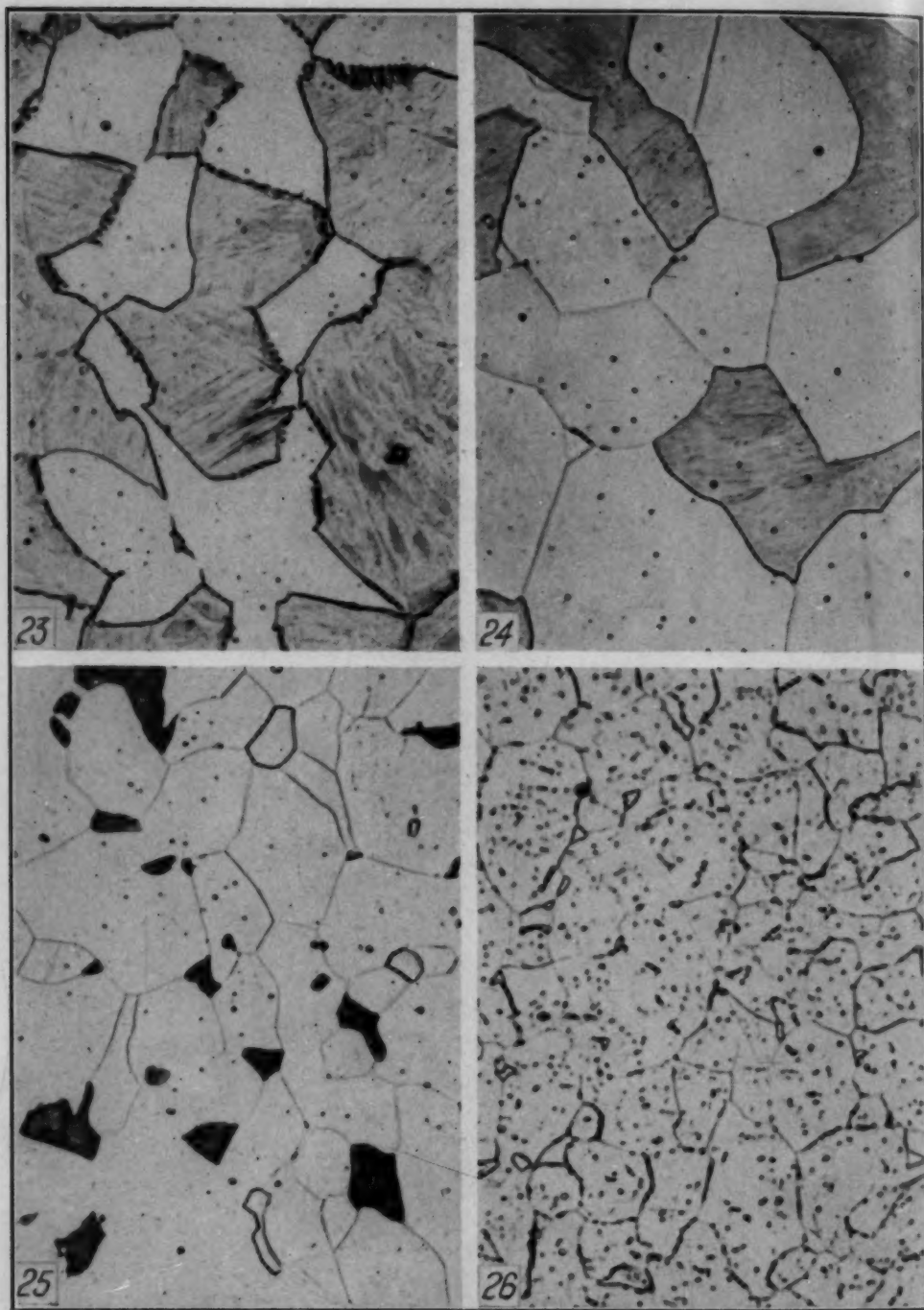
Fig. 23—Carbon 0.12 Per Cent, Molybdenum 0.008 Per Cent. Held 90,000 Seconds at 1500 Degrees Fahr. 37 Per Cent Transformed. × 1000.
Fig. 24—Carbon 0.11 Per Cent, Molybdenum 0.74 Per Cent. Held 90,000 Seconds at 1500 Degrees Fahr. 77 Per Cent Transformed. × 1000.
Fig. 25—Carbon 0.12 Per Cent, Molybdenum 0.008 Per Cent. Held 80,000 Seconds at 1300 Degrees Fahr. 93 Per Cent Transformed. × 1000.
Fig. 26—Carbon 0.11 Per Cent, Molybdenum 0.74 Per Cent. Held 80,000 Seconds at 1300 Degrees Fahr. Completely Transformed. × 1000.
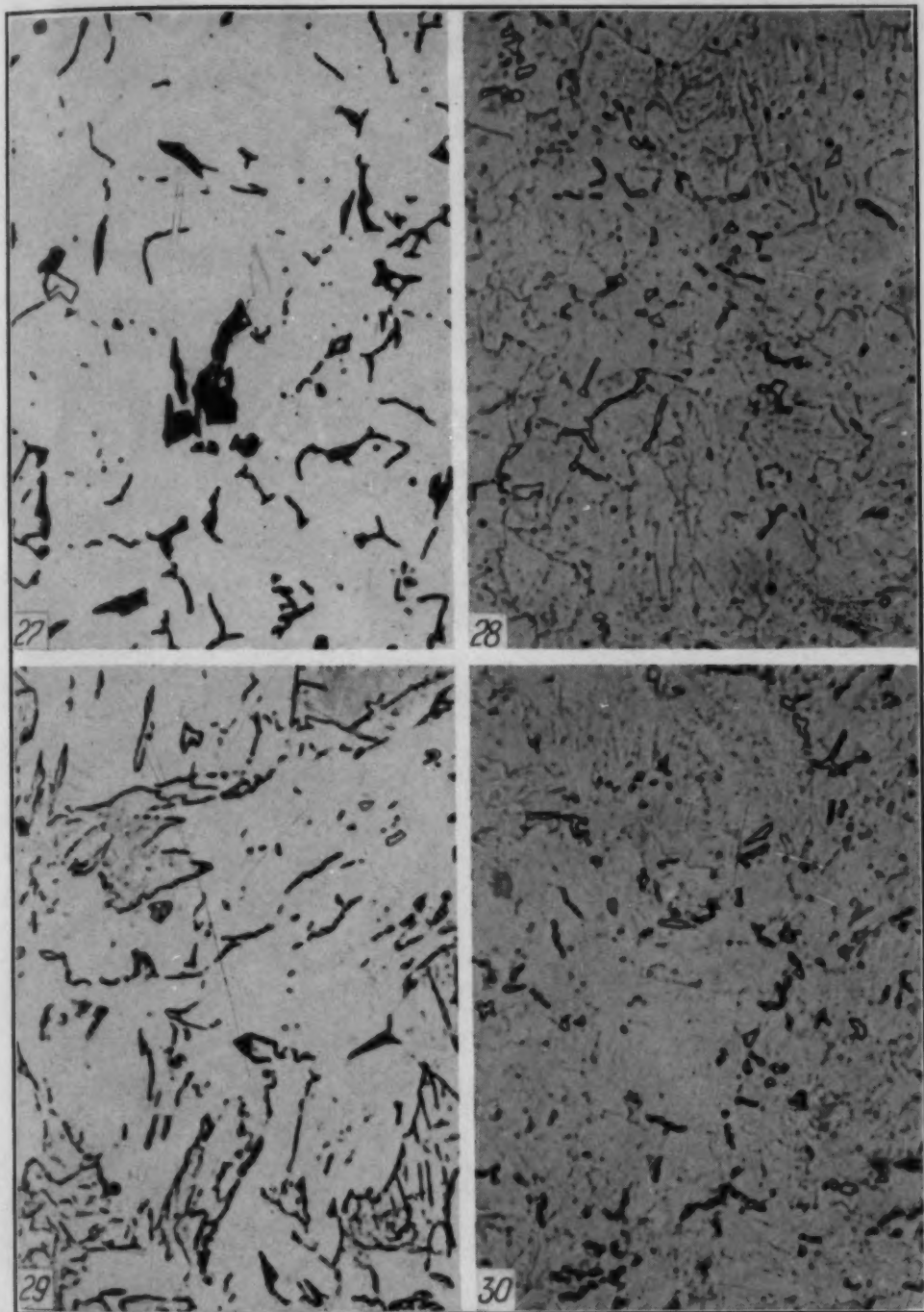
Fig. 27—Carbon 0.12 Per Cent, Molybdenum 0.008 Per Cent. Held 30 Seconds at 1200 Degrees Fahr. Completely Transformed. × 1000.

Fig. 28—Carbon 0.11 Per Cent, Molybdenum 0.74 Per Cent. Held 90,000 Seconds at 1200 Degrees Fahr. Completely Transformed. × 1000.

Fig. 29—Carbon 0.12 Per Cent, Molybdenum 0.008 Per Cent. Held 10 Seconds at 1100 Degrees Fahr. Completely Transformed. × 1000.

Fig. 30—Carbon 0.11 Per Cent, Molybdenum 0.74 Per Cent. Held 90,000 Seconds at 1100 Degrees Fahr. 98 Per Cent Transformed. × 1000.
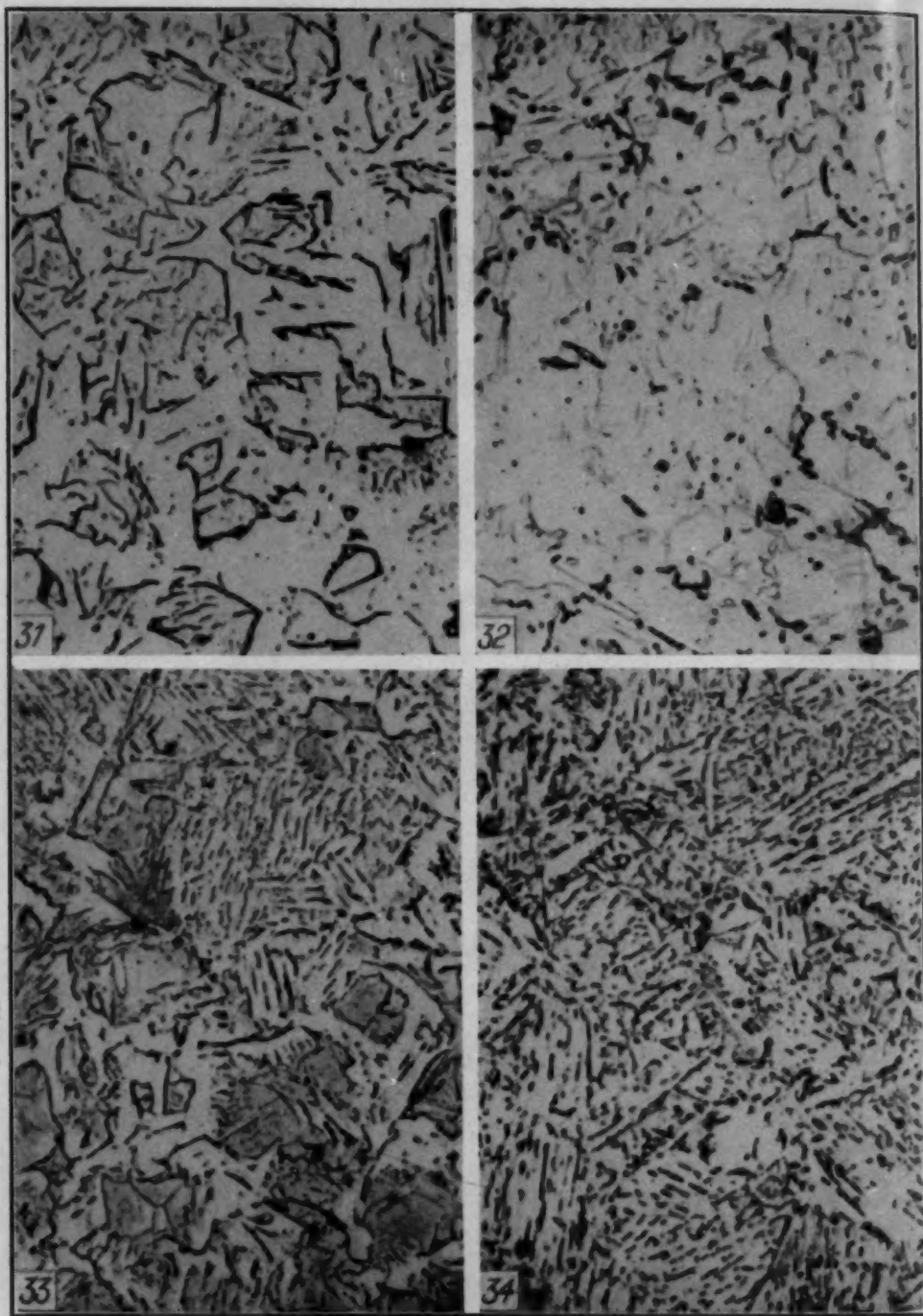
Fig. 31—Carbon 0.12 Per Cent, Molybdenum 0.008 Per Cent. Held 3 Seconds at 1000 Degrees Fahr. Completely Transformed. × 1000.

Fig. 32—Carbon 0.11 Per Cent, Molybdenum 0.74 Per Cent. Held 90,000 Seconds at 1000 Degrees Fahr. Completely Transformed. × 1000.

Fig. 33—Carbon 0.12 Per Cent, Molybdenum 0.008 Per Cent. Held 5 Seconds at 900 Degrees Fahr. Completely Transformed. × 1000.

Fig. 34—Carbon 0.11 Per Cent, Molybdenum 0.74 Per Cent. Held 10 Seconds at 900 Degrees Fahr. Completely Transformed. × 1000.
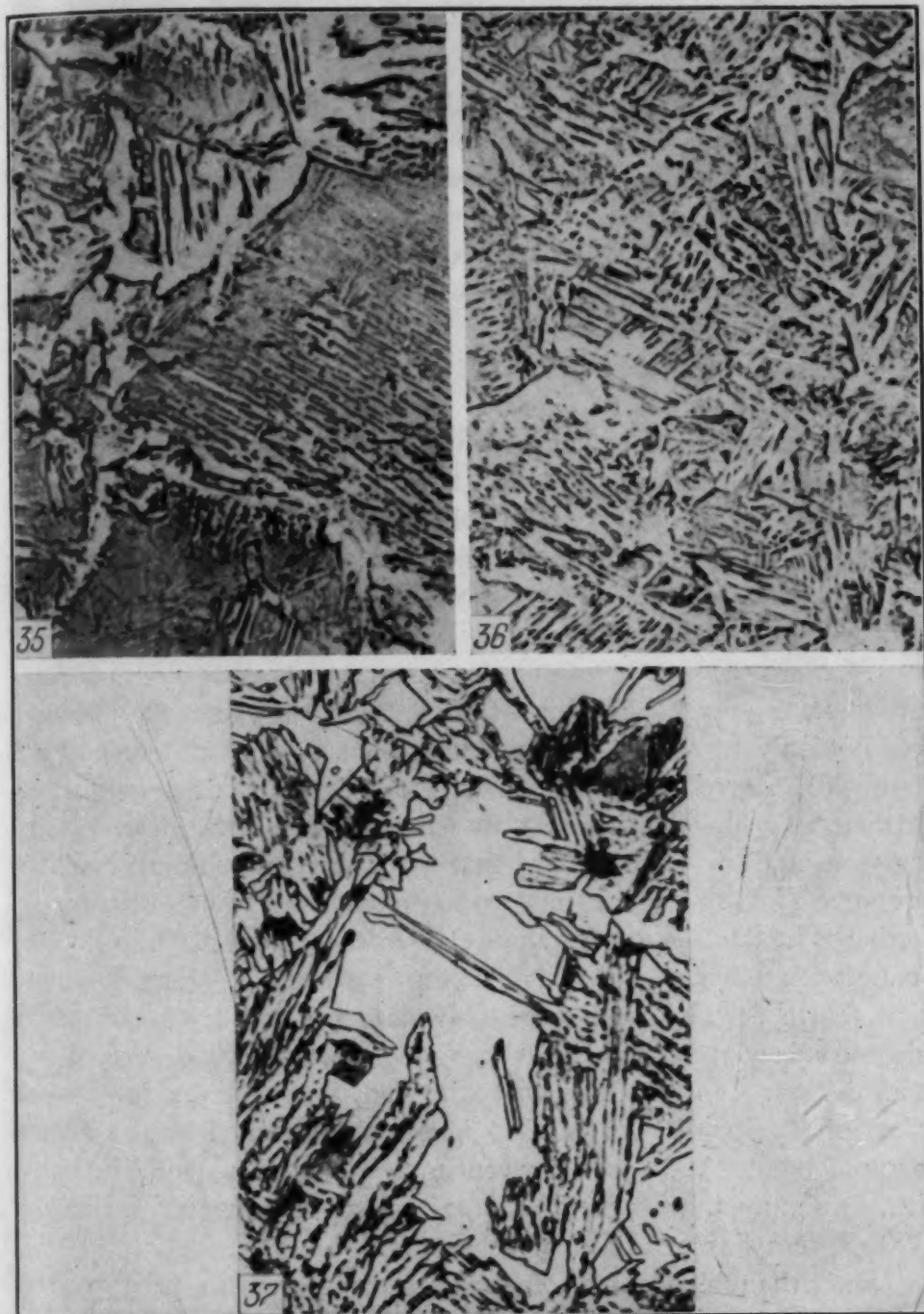
Fig. 35—Carbon 0.12 Per Cent, Molybdenum 0.008 Per Cent.　Held 5 Seconds at 700 Degrees Fahr.　Completely Transformed.　× 1000.
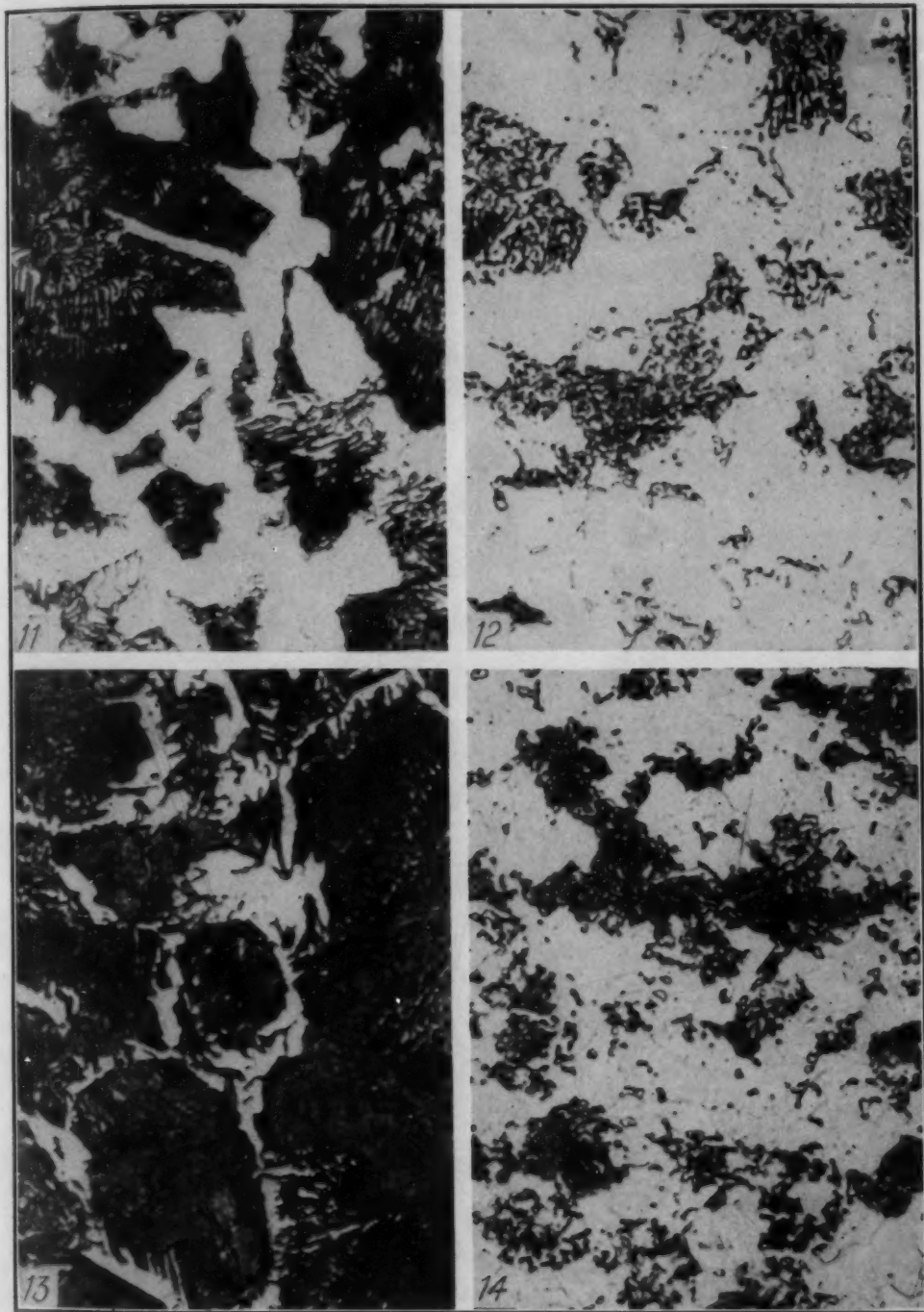Fig. 36—Carbon 0.11 Per Cent, Molybdenum 0.74 Per Cent.　Held 5 Seconds at 700 Degrees Fahr.　Completely Transformed.　× 1000.
Fig. 37—Carbon 0.39 Per Cent, Molybdenum 0.49 Per Cent.　Held 10 Seconds at 950 Degrees Fahr.　74 Per Cent Transformed.　× 1500.

Discussion of Results

Low Carbon Steels—The S-curves of the 0.12 per cent carbon steels (Figs. 1, 2, 3, 4, 5) show that the time for initiation of austenite transformation is only moderately affected by the addition of molybdenum up to 0.75 per cent. The rate of reaction at 1300 degrees Fahr. (705 degrees Cent.) is increased somewhat, but at temperatures in the range 900 to 1200 degrees Fahr. (485 to 650 degrees Cent.) the reaction rate is markedly decreased with increase in molybdenum concentration.

At temperatures below 900 degrees Fahr. (485 degrees Cent.) the reaction is extremely rapid in the low carbon series. In many cases only the finish of transformation was observed. It did not appear that molybdenum in amounts up to 0.75 per cent affected the reaction rate below 900 degrees Fahr. (485 degrees Cent.), although this statement might require modification since it cannot be assumed that the methods used here are capable of following a reaction that transpires in less than two seconds.

Medium Carbon Steels—The S-curves of the 0.42 per cent carbon steels presented in Figs. 6, 7, 8, 9, 10 show that molybdenum acts not only to delay the conversion of austenite at temperatures above 900 degrees Fahr. (485 degrees Cent.) but modifies the characteristic shape of the S-curve for carbon steels. The rate of transformation is so influenced that a secondary maximum reaction-rate zone (i.e., minimum time for completion of transformation) is produced in the S-curve at about 1200 degrees Fahr. (650 degrees Cent.) with 0.29 per cent molybdenum (see Fig. 8). Figs. 9 and 10 show that a further increase in molybdenum concentration displaces this secondary maximum reaction-rate zone to 1250 degrees Fahr. (675 degrees Cent.). It is to be noticed, too, that the ferrite line (start of precipitation of ferrite) is markedly altered by the molybdenum content. With 0.75 per cent molybdenum a secondary "nose" (i.e., minimum time for initiation of transformation) is formed at 1300 degrees Fahr. (705 degrees Cent.).

Since the molybdenum steels do not transform at a progressively increasing rate with decrease in transformation temperature in the range 1350 to 900 degrees Fahr. (735 to 485 degrees Cent.) as does the plain carbon steel, there are certain structures and hardnesses which are not likely to be obtained when these steels are continuously cooled at rates prevailing in such commercial treatments as an air cool or furnace cool.

The effect of a secondary nose on the response of a steel to slow cooling cycles is similar in nature to the effect of the principal nose at 1000 degrees Fahr. (540 degrees Cent.), the position of which determines the critical cooling rate of the steel.

The principal nose of the S-curve, which is fairly pointed in the plain carbon steel, is made more blunt with each addition of molybdenum.

It is informative to note that the reaction rate[8] just above the nose of the S-curve of the plain carbon steel is almost a maximum while in the molybdenum steels (i.e., with molybdenum greater than 0.30 per cent) the rate of transformation at the upper part of the nose is almost a minimum. This observation suggests that when the plain carbon steel is cooled at a rate just less than the critical cooling rate the product of transformation will be predominately lamellar, but when the molybdenum steels are cooled at rates just under their critical rate, the product of transformation is largely acicular.

At temperatures below 900 degrees Fahr. (485 degrees Cent.) the course of the S-curve is only slightly shifted to the right by molybdenum.

Below 500 degrees Fahr. (260 degrees Cent.) the transformation of austenite in the 0.42 per cent carbon series is divided into two stages. This phenomenon has been discussed by others (4), (5); so it was considered of interest to investigate the possibility that segregation is a cause of the interruption of the reaction. Samples from steels Nos. 6 and 10 were therefore homogenized by holding forty-eight hours at 2275 degrees Fahr. (1245 degrees Cent.) in vacuo. It was found, however, that the time versus dilation curves of these two homogenized steels at 400 degrees Fahr. (205 degrees Cent.) did not differ significantly from those curves made on the steels prepared in the usual manner. It appears, therefore, that chemical inhomogeneity is not the cause of the rate changes during transformation at temperatures below 500 degrees Fahr. (260 degrees Cent.).

The S-curves of the 0.42 per cent carbon steels are similar in trend to those presented for carbon-molybdenum steels by Davenport (2).

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[8] The S-curve does not expressly state rates of reaction, but the difference in time between initiation and completion of transformation may be taken as a measure of the average rate of transformation for any particular temperature level.

MICROSTRUCTURES

The microstructures of the products of isothermal austenite decomposition are variously modified by molybdenum. This is best described by referring to the photomicrographs in Figs. 11 to 36 where characteristic structures of the carbon steels are compared with those obtained at the same temperature when 0.75 per cent molybdenum is present.

The major effect of molybdenum on the microstructure of the 0.42 per cent carbon steel is in the volume increase and the character of the proeutectoid ferrite. From 950 to 1250 degrees Fahr. (510 to 675 degrees Cent.) the ferrite in the carbon steel appears almost wholly at the grain boundaries. Upon the addition of as little as 0.15 per cent molybdenum the formation of pearlite, in this same range, is greatly delayed so that there is a significant increase in the volume of ferrite produced, which thereupon precipitates both at and within the prior austenite grain boundary. This effect increases with increase of molybdenum up to at least 0.75 per cent.

In the medium carbon steels the tendency for the ferrite to precipitate in an acicular manner in the range 900 to 1100 degrees Fahr. (480 to 595 degrees Cent.) is promoted by molybdenum. This acicular type ferrite occurs in parallel blades which become more closely spaced with the decrease in transformation temperature. The pearlite which forms after the ferrite blades have grown in all directions occurs in smaller and smaller patches as the temperature of transformation is lowered. At approximately 950 degrees Fahr. (510 degrees Cent.) the character of the ferrite formation is such that the microstructure of the higher molybdenum steels consists chiefly of carbide-containing blades widely dispersed in ferrite. The hardness of this structure, however, is comparable to that of the carbon steel transformed at the same temperature. The carbide constituent apparently forms at the junction of the lateral growth of two ferrite blades as seen in Fig. 37. The final structure obtained then in the range 900 to 1100 degrees Fahr. (485 to 595 degrees Cent.) is a Widmanstätten type. The spacing between the carbide-containing constituents of this structure decreases with the lowering of the transformation temperature.

The Widmanstätten type structure has been associated with molybdenum steels for some time. It is evident from the S-curves

that this structure is likely to be encountered in molybdenum steels, since molybdenum retards the transformation rate at temperatures above those in which Widmanstätten is produced without delaying to a corresponding degree transformation in the Widmanstätten temperature range.

The 0.12 per cent carbon steels contain so small an amount of carbon that there is little opportunity for lamellar pearlite to form. In the range 900 to 1200 degrees Fahr. (485 to 650 degrees Cent.) molybdenum modifies the microstructure of the low carbon steel chiefly by retarding the transformation so that when carbide does precipitate it is less rod-like and more spheroidal in form.

At temperatures below 900 degrees Fahr. (485 degrees Cent.) the austenite decomposition products of the molybdenum steels are similar to those of the corresponding carbon steel in both carbon ranges.

CONCLUSIONS

A study of the isothermal, subcritical transformation of austenite in low and medium carbon steels containing molybdenum in amounts up to 0.75 per cent discloses the effectiveness of this element in promoting the hardenability of steel. In addition the distortions which molybdenum produces in the S-curves prepared from the isothermal data clearly indicate that the complete description of the influence of molybdenum on the decomposition of austenite is complex.

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5. J. B. Austin and R. L. Rickett, "Kinetics of the Decomposition of Austenite at Constant Temperature", *Transactions,* American Institute of Mining and Metallurgical Engineers, Vol. 135, 1939, p. 396.

DISCUSSION

Written Discussion: By H. H. Chiswik, Teaching Fellow, Laboratory of Physical Metallurgy, Harvard University, Cambridge, Mass.

In the course of an investigation, as yet unpublished, dealing with the mechanism and Ar″ temperatures of the martensite transformation in alloy steels, we had occasion to use an alloy of carbon content equivalent to that used by the authors (0.40 per cent), but containing 1.5 per cent molybdenum. All our work was microscopic and the experimental technique employed was that described by Greninger and Troiano.[4] The Ar″ temperature for this alloy was found to be 375 degrees Cent. (± 5 degrees Cent.) (707 degrees Fahr.). No direct comparison can be made of this value with the Ar″ temperature of a plain 0.40 per cent carbon steel, since no accurate experimental data are available for steels with carbon contents lower than 0.60 per cent. However, it is near the estimated value of 350 degrees Cent. (662 degrees Fahr.), obtained by extrapolation from higher carbon values,[4] assuming the Ar″ to vary linearly with carbon content. These results, therefore, seem to indicate that small percentages of molybdenum have little effect, if any, on the position of the Ar″ temperature. At 400 degrees Cent. (752 degrees Fahr.), austenite of this alloy begins to decompose after 7 to 8 seconds at temperature, indicating a definite retardation in the beginning of austenite decomposition at temperatures above the Ar″, when compared with a plain 0.40 per cent carbon steel. Furthermore, molybdenum was found to have no effect on the mechanism or kinetics of martensite formation when compared with that described by Greninger and Troiano[4] for plain carbon steels.

A careful examination of the "Reaction begins" curves, as given by the authors in Figs. 7, 8, 9 and 10, reveals that the curves reverse direction at one constant temperature, 700 degrees Fahr. (371 degrees Cent.), regardless of the molybdenum content. This suggests a unique property for this temperature, which agrees surprisingly well with our experimentally determined value of the Ar″ temperature for a steel of similar carbon content. It is, therefore, suggested that this is the Ar″ temperature. At all temperatures above 700 degrees Fahr. (371 degrees Cent.)—the Ar″ temperature—the curves indicate that molybdenum retards the beginning of austenite decomposition when compared with plain carbon steels of similar carbon content, which agrees well with our findings. At temperatures below 700 degrees Fahr. (371 degrees Cent.), however, the picture becomes more complicated. Since the Ar″ cannot be suppressed by quenching, all samples quenched to below 700 degrees Fahr. (371 degrees Cent.) are no longer pure austenite the instant they reach the temperature of the quenching bath. Thus, a sample quenched into a 500 degrees Fahr. (260 degrees Cent.) bath and held there for one second, contains about 75 per cent martensite at that temperature, judging from our observations. Holding, therefore, a sample for some time at any temperature below 700 degrees Fahr. (371 degrees Cent.) will undoubtedly cause both the remaining austenite and the martensite produced during the quench to decompose at their respective rates. Such a state of affairs would account for the

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[4] A. B. Greninger and A. R. Troiano, "Kinetics of the Austenite to Martensite Transformation in Steel," TRANSACTIONS, American Society for Metals, Vol. 28, 1940, p. 537.

distinct discontinuity or irregularity in the reaction rate observed dilatometrically at about 500 degrees Fahr. (260 degrees Cent.).

It should be emphasized here, that one cannot measure the rate of austenite decomposition at constant temperature below the Ar″, unless he has a means of distinguishing between martensite and the constant temperature decomposition product of austenite. The microscopic method is satisfactory at temperatures just below the Ar″, where the amount of remaining austenite is large. Dilation methods are inapplicable, since they offer no means of distinguishing between the two products, and unless temperature is recorded simultaneously with dilation, they may give completely erroneous indications of the beginning of decomposition.[5] They further complicate the picture by recording the phenomena accompanying the decomposition of martensite.

Oral Discussion

E. S. DAVENPORT:[6] Naturally, we have found this paper extremely interesting, as is the case with all of the isothermal transformation work emanating from the Climax Molybdenum Company laboratory; it is exceedingly gratifying to see independent confirmation of some of these unusual diagrams which, when we first began to find them a few years ago we hardly believed ourselves. I see Mr. Rutherford here in the audience. He took an active part in some of that early work, and I think he will confirm my statements when I say that when we first began to develop these alloy steel diagrams, we thought something had gone wrong with our experimental technique. We checked ourselves repeatedly, however, and now that the present authors are finding the same general trends we feel somewhat comforted.

I would like to make one suggestion: We find it quite useful, in applying these diagrams, to have the A_3 and the A_1 line indicated on them; it provides a sort of "roof" for the top of the diagram to which the curves may be drawn asymptotically, as theory and observations indicate they should be drawn. I think it might be helpful to insert these lines in the diagrams if the information is available.

One of the important items of information lacking at the present time, has to do with the mechanical properties of the products of isothermal transformation in alloy steels. We have not had an opportunity to obtain this information, although English investigators have studied the mechanical properties of some of these isothermally transformed alloy steels; I wonder if the authors have started work on this phase of the subject?

J. J. B. RUTHERFORD:[7] I wonder if the authors of this paper have noted any general relationship between the behavior of the carbide-forming elements, such as molybdenum, and the transformation products which have developed, and the relationship between the products which have formed and the type of the S-curve.

[5]N. P. Allen, L. B. Pfeil and W. T. Griffiths, "The Determination of the Transformation Characteristics of Alloys Steels," The Iron and Steel Institute, Second Alloy Steels Report, 1939, p. 369.

[6]Metallurgist, United States Steel Corp., Kearny, N. J.

[7]Metallurgist, The Babcock and Wilcox Tube Co., Beaver Falls, Pa.

More specifically what I have in mind is this: In the carbon steels we have two general types of reaction, austenite-ferrite transformations to the lamellar products at the high temperature and the transformation to the acicular products at the lower temperature, both of these reactions involving merely transformation from a solid solution to a ferrite-iron carbide mixture.

When we get into the alloys of the type that have been described here, we encounter mixtures of ferrite, iron carbide, and various compounds of molybdenum carbide which naturally are bound to have effects on these reactions.

Authors' Reply

The authors wish to thank the gentlemen who discussed this paper for their remarks. We agree with Mr. Davenport's suggestion that the A_3 and A_1 temperature lines would be a helpful addition to the diagrams, and we intend to include such lines in the future.

We have no extensive information on the mechanical properties of the isothermal transformation products of molybdenum steels. We are continuing the isothermal, subcritical transformation studies in a current investigation of molybdenum steels with 0.80 per cent and 1.20 per cent carbon. We feel that these fundamental data are more urgently needed than the mechanical property data, but we hope to investigate the mechanical properties in the future.

In answer to Mr. Rutherford's question concerning complex carbides, we have been unable to detect complex molybdenum carbides in the microstructures of these steels.

Thanks are extended to Mr. Chiswik for his confirmation of a portion of our data. His comments on the highly controversial subject, the Ar'' temperature, are appreciated. The authors are well aware of the fact that the dilatometer does not discriminate between the small contraction due to the tempering of transformed material and the relatively large concomitant expansion of the gamma-alpha change. For determination of the beginning and ending of transformation, this error, at least in steels of 0.40 per cent carbon or less, is not significant since agreement between the metallographic and dilatometric methods is good. Were the data to be used in *rate* studies, however, it would be desirable to take account of the influence of secondary decomposition reactions on the time-dilation curves.

The dilatometer used in this investigation indicated the specimen length, not only during isothermal transformation, but also during cooling from the supercritical temperature to the subcritical temperature. It was, therefore, possible to ascertain when the rate of transformation relative to the rate of cooling became so fast that the beginning of transformation was indeterminate. Thus, when transformation was found to start within two seconds, the beginning line of the S-curve was dotted.

CEMENTITE STABILITY AND ITS RELATION TO GRAIN SIZE, ABNORMALITY AND HARDENABILITY

By CHARLES R. AUSTIN AND M. C. FETZER

Abstract

From a study of the behavior of eighteen similar hypereutectoid steels it has been suggested in a previous paper that the presence of alumina may serve as a graphitizer during prolonged tempering. In the present paper the authors pursue further this study on the steels and report on the correlation between carbide stability or tendency to graphitization and fracture grain size at various temperatures, McQuaid-Ehn grain size and abnormality, and hardenability of the steels.

The tempering studies were made in covered lead at 670 degrees Cent. for periods up to 600 hours, on specimens quenched from 1000 degrees Cent. Ten steels exhibited complete resistance to graphitization, whereas eight of the steels showed graphitization ranging from approximately 50 to 100 per cent.

In general, the results indicated that coarse-grained normal steels were stable while the fine-grained abnormal steels graphitized. Evidence has been provided to show that alumina, which may have such a profound effect on the control of grain size and abnormality may similarly markedly lower the stability of the carbide. The data on hardenability obtained from the various steels at common grain size did not permit any attempt at correlation with carbide stability. The relative deep hardening of a few of the steels has been attributed to the slightly higher silicon plus manganese content.

I T was suggested in a recent paper[1] by the authors that the graphitization of hypereutectoid carbon steels at subcritical temperatures was due to the presence of certain minute refractory deoxidation products in the steel. Alumina, formed during the usual commercial deoxidation of the molten steel or by the oxidation of

metallic aluminum during heat treatment, appeared to play the major role. If such is the case, then the most important factor controlling graphitization of these steels is the same one that functions so largely in controlling grain size, abnormality, and hardenability. A correlation of graphitizing tendency with these other properties might then be expected. Such a relationship would not only lend strong support to the proposed theory of graphitization but might also aid in the selection or rejection of steels where graphitization or carbide stability may be important.

Accordingly, tests were conducted on the eighteen steels considered in the paper cited,[1] with a view to providing an experimental study of the above inferences.

Table I

Composition—Per Cent of the 18 Steels

Steel	C	Si	Mn	S	P	Cr	Ni	Cu	N	Al	Al$_2$O$_3$	Total Oxygen
H	1.08	0.33	0.36	0.017	0.020	0.11	0.04	0.12	0.0065	0.010	0.006	0.005
I	1.02	0.27	0.22	0.010	0.014	0.09	0.04	0.13	0.0070	0.005	0.004	0.007
J	1.03	0.27	0.31	0.020	0.017	0.10	0.04	0.10	0.0045	none	0.002	0.005
K	1.00	0.45	0.30	0.022	0.020	0.06	0.06	0.13	0.0070	none	0.001	0.013
L	1.07	0.31	0.26	0.010	0.020	0.09	0.06	0.15	0.0059	0.003	0.004	0.008
M	1.00	0.20	0.18	0.021	0.017	0.08	0.04	0.17	0.0093	0.010	0.007	0.005
N	1.12	0.20	0.25	0.013	0.017	0.04	0.13	0.13	0.0101	0.007	0.008	0.004
O	1.06	0.13	0.19	0.010	0.016	0.03	0.06	0.13	0.0097	0.010	0.008	0.0055
P	1.07	0.14	0.22	0.010	0.017	0.07	0.07	0.15	0.0101	0.003	0.006	0.0045
R	1.01	0.28	0.34	0.021	0.010	0.06	0.10	0.08	0.0034	0.001	0.004	0.004
S	1.03	0.21	0.24	0.017	0.017	0.04	0.04	0.15	0.0104	0.004	0.006	0.005
T	1.01	0.21	0.21	0.019	0.012	0.10	0.08	0.13	0.0042	0.003	0.007	0.0035
U	1.02	0.20	0.27	0.016	0.013	0.11	0.10	0.16	0.0042	0.020	0.009	0.0035
V	1.03	0.20	0.20	0.023	0.013	0.11	0.06	0.13	0.0034	0.002	0.008	0.0035
W	1.03	0.21	0.18	0.020	0.010	0.09	0.03	0.19	0.0039	0.018	0.010	0.0045
X	0.99	0.24	0.22	0.017	0.015	0.04	0.02	0.11	0.0084	0.019	0.011	0.0055
Y	1.03	0.22	0.20	0.014	0.012	0.05	0.02	0.06	0.0031	0.004	0.009	0.0045
Z	0.98	0.19	0.20	0.017	0.014	0.04	0.04	0.04	0.0014	0.004	0.010	0.007

The chemical analyses of the eighteen hypereutectoid carbon steels are given in Table I and it may be noted that the variation in amounts of impurities is, with few exceptions, such as to permit their classification under "steels of similar chemical composition". Furthermore, in general, the small differences noted are usually considered as insufficient to render markedly different the behavior of these steels in their reaction to tempering. The special analyses in Table I are included merely for completeness; only those which appear to have a bearing on the results obtained will be discussed. Additional data are given in Table III; namely, the manufacturer, the type of furnace used, and the aluminum additions to furnace, ladle, and mould.

Present Investigation

Suitable heat treatment was given to the steels in order to study the relative carbide stability or tendency toward graphitization, and its correlation with fracture grain size, McQuaid-Ehn grain size and abnormality, and hardenability as indicated by the Jominy end-quench method.

1. *Carbide Stability or Graphitization*—Since there is evidence to indicate that differences in tendency to graphitize may be modified by the atmosphere surrounding the steels during heat treatment, and by the nature of the metallographic structure prior to the graphitizing treatment, attention must be given to these factors. With regard to the effect of atmosphere all heat treatments were conducted so that no scaling or evidence of surface decarburization was observed even after 600 hours at 670 degrees Cent. (1240 degrees Fahr.). This condition was met by (a) annealing in charcoal for 1 hour at 1000 degrees Cent. (1830 degrees Fahr.) followed by water quenching, and (b) heating at 670 degrees Cent. (1240 degrees Fahr.) for 600 hours in lead, protected with charcoal, followed by air cooling.

It may be assumed that this treatment is not likely to modify, to any marked extent, what is termed the "inherent" carbide stability or tendency toward graphite formation in the steels. If the steels are not quenched after the high temperature pretreatment the rate of graphitization is markedly slower. The relative degree of graphitization was indicated by the hardness of the steel observed at the end of the test, and a check on the data was obtained from a study of the microstructure.

The hardness data resulting from the above heat treatments are given in Table II.

Table II

Rockwell "B" Hardness of the Eighteen Steels After 600 Hours at 670 Degrees Cent. in Lead Protected by Charcoal. Pretreatment: 1 Hour at 1000 Degrees Cent., and Water Quenched

Steel	R_B	Steel	R_B
H	88	[S]	50
I	86	[T]	48
J	87	[U]	61
K	87	[V]	46
L	87	[W]	43
M	86	[X]	42
N	86	[Y]	46
O	86	[Z]	38
P	86		
R	86		

Steels in brackets graphitize in covered lead at 670 degrees Cent.

The ten steels, H to R, listed in column 1 of Table II did not graphitize and the completely spheroidized structure gave a Rockwell B hardness value within the limits 86 and 88. Fig. 1 shows at 1000 diameters the structure of steel H which is typical of the steels having stable cementite. Complete graphitization of the steels results in a hardness value of about 35 B, whereas 50 per cent graphitization after 600 hours tempering, is indicated by approximately 60 B. It is to be noted, therefore, that the eight steels listed in column 3, and printed in brackets, are graphitized in amounts varying from about 50 to 100 per cent. Fig. 2 shows a field from an etched section
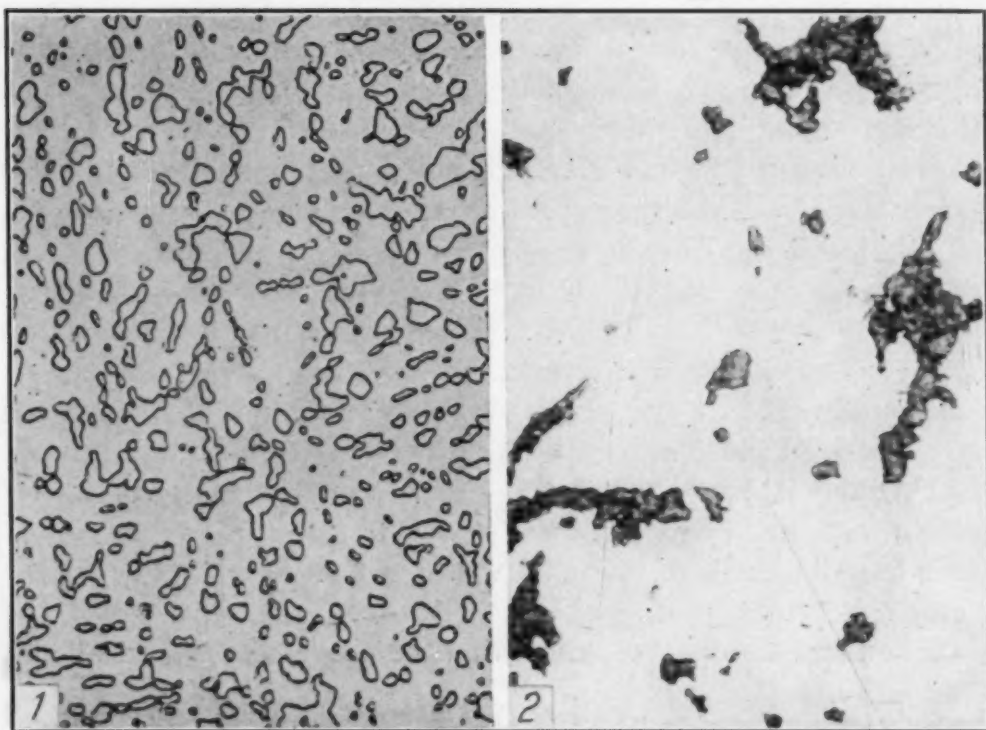
Fig. 1—Spheroidized Cementite, Steel H, 600 Hours at 670 Degrees Cent. in Covered Lead. Pretreatment 1 Hour at 1000 Degrees Cent. in Charcoal and Water Quenched. Hardness—88B. Etched. × 1000.
Fig. 2—Retained Graphite in Ferrite, Steel Z, Heat Treatment Same as Fig. 1. Hardness—38B. Etched. × 1000.

of specimen Z at 1000 diameters where graphitization was practically complete. This picture shows more graphite than the average content of the steel but it indicates the true area occupied by the graphite at this location as is evidenced by the gray color and sharp outline of the irregular areas. The polishing technique used to retain the graphite particles was as follows: A mixture of magnesia and paraffin was prepared, and rubbed into a felt cloth for the final pol-

ishing operation. By alternately etching and polishing several times and by using water on the wheel to prevent tarnishing the graphite remains as a blue-gray clearly defined phase.

A study of the correlation between carbide stability, as indicated by the hardness data, and the aluminum and alumina chemical analyses given in Table I, or the aluminum additions made in manufacture (Table III) permit the following observations:

1. With the graphitizing steels (no data for X, Y, and Z) the total amount of aluminum added to the molten metal amounted to approximately 1 pound per ton or more. Aluminum additions to the stable steels were less than 1 pound per ton in all cases.

2. No aluminum addition to the mould was made in the manufacture of the ten stable steels but such addition was made to the graphitizing steels S and V, the only two unstable steels on which any information regarding aluminum additions to the mould was available.

3. The steels analyzing 0.004 per cent alumina and less were stable while those with over 0.008 per cent were unstable. The steels containing intermediate amounts, 0.006 to 0.008 per cent fell in either the stable or graphitizing group. The aluminum and alumina analyses are relatively high for steels X, Y, and Z, on which there were no melting practice data available, indicating that these three graphitizing steels also had quantities of aluminum added to the melt.

Observations (1) and (3) indicate that all except three of the steels, K, L, and R, contain alumina and that 15 of the 18 steels should graphitize. From observation (2), however, it may be concluded that the mere presence of alumina does not suffice but that the form or degree of dispersion of the alumina is important.

2. *Austenitic (Fracture) Grain Size*—The grain size characteristics of the eighteen steels were determined as follows: The steels were first heated for 1 hour at 900 degrees Cent. (1650 degrees Fahr.) followed by oil quenching. Samples of these pretreated materials were then subjected to one-half hour at one of the following temperatures and water-quenched: 790, 845, 900, 955 and 1010 degrees Cent. (1450, 1550, 1650, 1750, and 1850 degrees Fahr.). Fracture grain size numbers were then obtained on all samples. The measurements made in the laboratory, and confirmed by the co-operation of G. V. Luerssen, are listed in Fig. 3 where an attempt has been made to show the relation between quenching temperature and fracture grain size, and to indicate whether the curve for a given

Table III

Classification of the 18 Steels on the Basis of McQuaid-Ehn Grain Size and Abnormality

Steel*	McQuaid-Ehn Grain Size		Normality		Aluminum Additions			Lbs./t. Total	Made by Co.	Kind of Furnace Made in**
	Case G. S.	Core G. S.	Case	Core	Furnace	Ladle	Mould			
(A) Fine in Case and Core										
M	7-8	7-8	Abn.	Abn.	none	0.50	none	0.50	B	B. A. E.
[S]	7-8	No net.	Abn.	....	1.10	0.41	0.14	1.65	B	B. A. E.
[Y]	6-8	6-8	Abn.	Abn.	....	....	....	....	C	A. E.
(B) Fine in Case—Mixed in Core										
[V]	7	50% 2-3; 50% 5-6	Abn.	Mixed	0.83	0.42	0.14	1.39	B	B. A. E.
[T]	7	80% 2-4; 20% 5-6	Abn.	Mixed	none	0.93	....	0.93	B	B. O. H.
[U]	7	60% 1; 40% 5-6	Abn.	Mixed	none	1.05	....	1.05	B	B. O. H.
[W]	7	75% 1-2; 25% 4-5	Abn.	Mixed	none	0.96	....	0.96	B	B. O. H.
(C) Mixed in Case and Core										
[X]	1-5	1-5	Mixed	Mixed	....	....	....	....	C	A. E.
H	2-6	2-6	Mixed	Mixed	0.11	0.22	none	0.33	A	B. A. E.
N	60% 2-3; 40% 4-5	60% 2-3; 40% 4-5	Mixed	Mixed	0.66	0.22	none	0.88	A	B. A. E.
P	75% 2-3; 25% 4-6	75% 2-3; 25% 4-6	Norm.	Norm.	0.66	0.22	none	0.88	A	B. A. E.
(D) Coarse in Case and Core										
R	4	4	Norm.	Norm.	none	none	none	none	D	H. F. I.
J	4	4	Sl. Abn.	Norm.	0.11	0.22	none	0.33	A	B. A. E.
I	3-4	3-4	Sl. Abn.	Norm.	0.11	0.22	none	0.33	A	B. A. E.
L	3-4	3-4	Norm.	Norm.	none	none	none	none	C	H. F. I.
[Z]	3-4	3-4	Sl. Abn.	Sl. Abn.	....	....	....	....	C	A. E.
(E) Mixed Coarse to Very Coarse										
O	1-3	50% 1-2; 50% 3-4	Mixed	Mixed	0.66	0.22	none	0.88	A	B. A. E.
K	1-4	1-3	Norm.	Norm.	none	none	none	none	A	H. F. I.

*Steels in brackets graphitize in covered lead at 670 degrees Cent.

**B. A. E. = Basic Arc Electric; A. E. = Arc Electric; H. F. I. = High Frequency Induction; B. O. H. = Basic Open-Hearth.

Fig. 3—Fracture Grain Size Curves.

steel belongs to the stable or to the graphitizing group of alloys. The results in terms of carbide stability are discussed later.

3. *McQuaid-Ehn Grain Size and Abnormality*—Samples were packed in "Char" compound, heated at 925 degrees Cent. (1700 degrees Fahr.) for 8 hours and then cooled to 650 degrees Cent. (1200 degrees Fahr.) at an average rate of 3.2 degrees Fahr. per minute. On polished and etched sections of the specimens grain size measure-

ments were obtained for the case, and also for the core where no increase in carbon content occurred. These results are found in Table III. Records were also made of the relative normality of the case and core and the observations are included in this table.

Except for steels M, S, and Y, which showed a very fine-grained core it was generally observed that the case was somewhat finer and more abnormal than the core. Thus when the steel showed a mixed grain in the core after treatment, the case was sometimes fine-grained. This refinement of the case is assumed to be accounted for
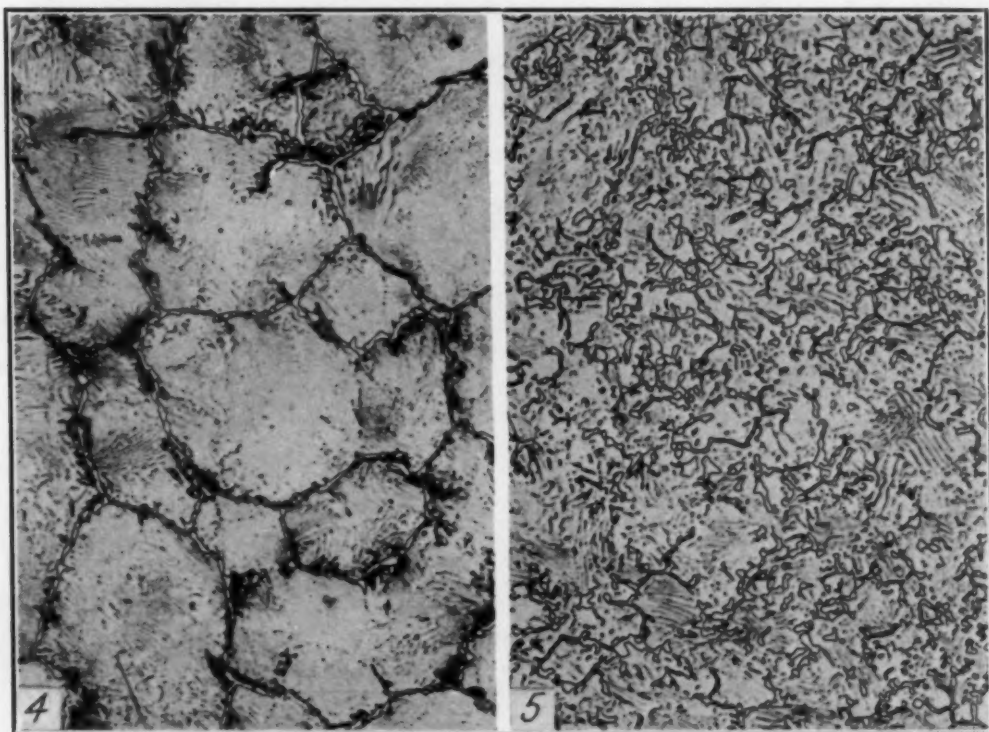
Fig. 4—Normal Structure at Carburized Case, Steel L. Electrolytic Sodium Picrate Etch. × 250.
Fig. 5—Abnormal Structure at Carburized Case, Steel S. Electrolytic Sodium Picrate Etch. × 250.

by oxygen penetration. On the basis of grain size in case and core a classification into five groups has been made as shown in Table III, which also lists data obtained from a study of the abnormality of the steels, as well as available information on melting practice. The relationship between these groups and graphitization is discussed later.

A typical example of the microstructure of the case of the normal steels is shown in the etched section of steel L, at 250 diameters, in Fig. 4. The abnormal condition of the case, observed in the micro-

structures of steels listed in groups A and B (Table III) is well shown in Fig. 5, representing the etched structure of steel S.

4. *End-Quench Hardability Tests*—In order to eliminate the effect of austenitic grain size, and to permit a study of the effect of the temperature from which the steels were quenched, on the hardenability properties, two series of tests "S" and "T" were conducted. Specimens for the two series were first quenched from 900 degrees Cent. (1650 degrees Fahr.) in oil and then chromium plated on one end, after cutting and polishing, to provide a non-scaling surface. Each of the steels in series "S" was then heated in a muffle for ½ hour at the temperature required to give a fracture grain size within the limits 4½ to 5½. The temperature required to obtain this common grain size was obtained from the data plotted in Fig. 3, and varied from 845 degrees Cent. (1550 degrees Fahr.) for the steel Z to 1010 degrees Cent. (1850 degrees Fahr.) for steels M, S, and Y. After the treatment each steel was transferred to a second furnace for 10 minutes at 845 degrees Cent. (1550 degrees Fahr.), and then end-quenched by the Jominy method. This procedure ensured uniform austenitic grain size and eliminated any effects which might have resulted due to differences of quenching temperature.

Each of the steels in series "T" was similarly treated except that the procedure of lowering to a common quenching temperature was omitted. They were quenched directly from the temperature required to give the 4½ to 5½ common grain size. These experiments afforded data for a study of the effect of T maximum or variation in quenching temperatures for a common austenitic grain size. The quenched samples were split longitudinally, polished and etched, and the depth of hardness measured to the region where the structure consisted of equal amounts of martensite and pearlite. The results are grouped in Table IV on the basis of hardness penetration. It is to be noted that there is little difference between the results of series S and series T, indicating that in the temperature range used, variation of the quenching temperature alone, has no appreciable effect on hardenability.

DISCUSSION OF RESULTS

The Ten Stable Steels H-R—Three of the steels, K, L, and R, had no aluminum added to the heat and it may be noted that these three steels show a coarser fracture than any of the other fifteen after treatment at temperatures over 870 degrees Cent. (1600 de-

Table IV
End Quench Hardenability Results

(Depth Measured to 50 Per Cent Pearlite and 50 Per Cent Martensite at $\times$ 40)
(Fracture Grain Size 4½ to 5½ in all Cases)*

Depth in Inches	S. Series† Quenched From 1550 Degrees Fahr.	T. Series† Quenched From G. S. Temp.
0.300	K	
0.295		K
0.290		
0.285		
0.280		
0.275		
0.270		
0.265		
0.260		
0.255		
0.250		
0.245		
0.240		
0.235	R	R
0.230		
0.225		
0.220		
0.215	J	
0.210		
0.205		J
0.200	H	
0.195		
0.190		
0.185		L
0.180	L I	
0.175		[X]
0.170	[S] [X]	I
0.165	[U]	[U]
0.160		
0.155	[Y]	O [S]
0.150	M O	[W] P
0.145	P N	[Y] N [V]
0.140	[Z]	[Z]

†Steels in brackets graphitize in covered lead.
*Several steels have been eliminated because they did not meet grain size requirement of 4½ to 5½ F.G.S.

grees Fahr.). They are also coarse-grained by the McQuaid-Ehn test and are definitely normal in both case and core. Incidentally these steels were made in induction furnaces and are comparatively high in silicon and manganese, while steel K is unusually high in silicon. None of these three steels graphitize when tempered in covered lead. The results then indicate that cementite is stable in steels which are coarse-grained, normal, and contain no alumina.

Three other steels, H, I, and J, had only a small amount of aluminum added to the melt, namely, 0.33 pounds per ton. The analyses, Table I, show that the aluminum and alumina are indeed low for steels I and J but are appreciably higher for steel H. From Fig. 3 it may be noted that steels I and J have become coarser at 1010 degrees Cent. (1850 degrees Fahr.) than any but the first three steels

discussed (K, L, and R). Steel H, however, has fine fracture grain characteristics. The McQuaid-Ehn tests confirm these observations, but reveal the mixed grain characteristic of both case and core in steel H. The coarse-grained steels are quite normal whereas mixed grain is accompanied by normal and abnormal microstructure. The conclusion to be drawn from the behavior of H, I, and J, inasmuch as none of these graphitize, is that with small aluminum additions carbide stability may be retained.

The remaining four steels in this group (M, N, O, and P) had slightly more aluminum added to the metal, i.e., 0.50 to 0.88 pounds per ton and all analyzed for alumina content between the limits 0.006 to 0.008 per cent. The effectiveness of these intermediate aluminum additions on grain size and abnormality is shown in Table III and Fig. 3. Mixed grain sizes and mixed normalities were found in three (N, O, and P) of the four steels. In steel M, apparently, prior deoxidation was so adjusted that small aluminum additions resulted in fine grain and abnormal McQuaid-Ehn structure. Despite the fact that the aluminum addition has rendered one of the steels abnormal and very fine-grained, all are resistant to graphitization. One factor to be noted, however, regarding all ten stable steels is that no aluminum has been added to the mould.

The Eight Graphitizing Steels—The steels which graphitized on tempering are S to Z included, listed in the second column of Table II, and printed in brackets in Table III, where it may be noted that five of the eight alloys (S, T, U, V, and W) had aluminum additions of 1 pound per ton or more. No information could be obtained from the manufacturers on the deoxidation practice used with steels X, Y and Z but chemical analysis showed steel X to have as high aluminum content, and all three steels to contain as much alumina as any of the alloys investigated.

The five steels which were reported to have aluminum additions, as well as steel Y, are fine-grained and abnormal in the carburized case, and show fine to medium fracture grain size (See Table III and Fig. 3). This leaves as exceptions steels X and Z. Steel Z is particularly interesting because it is not only rather coarse-grained and normal but also is the most readily graphitizable steel.

GENERAL CONCLUSIONS

The preceding observations lead to the following conclusions:

1. Steels to which no aluminum has been added and which for

this reason are coarse-grained and normal will not graphitize at subcritical temperatures.

2. In general, steels to which aluminum additions have been made and which show a fine-grained abnormal case in the McQuaid-Ehn test, will graphitize (exception—steel M).

3. With two exceptions (X and Z) steels to which aluminum has been added but which remain coarse (normal) or even mixed grained (mixed normality) in the McQuaid-Ehn test, do not graphitize.

The exceptions to the above generalizations are interesting and merit comment. The two steels to be discussed are M and Z. The former is fine-grained and abnormal yet is stable, whereas the latter is coarse-grained and normal, but graphitizes readily. Considering first steel Z, no deoxidation data could be obtained, but chemical analysis strongly supports the premise that aluminum was added for deoxidation purposes. Apparently this addition produced an alumina in the steel in form suitable for the promotion of graphitization but not for the control of grain size. However, it is well known that the addition of aluminum to steel, per se, does not necessarily result in a fine-grained alloy. The molten metal must first be prepared by suitable deoxidization. From the analyses it would seem that the steel was insufficiently deoxidized prior to the aluminum additions in order for that element to function as a controller of grain size.

With reference to the behavior of steel M, an example is furnished where the additions of aluminum provided alumina in sufficient quantity and in a suitable degree of dispersion to yield a fine-grained alloy. However, no evidence of carbide instability was observed after prolonged tempering at 670 degrees Cent. (1240 degrees Fahr.). The only correlation with the other nine stable steels, which may be pointed out, is that even though aluminum addition was made it was done prior to the casting in the mould. Furthermore this addition was limited to 0.50 pound per ton, and as suggested in the discussion on steel Z the nature of the dispersed alumina is important and may function differently in grain control as compared with graphitization.

Kolchin and Assur[2] in a recent study of two 1 per cent plain carbon strip steels, one of which was normal and the other abnormal, found that the normal steel was difficult to graphitize even at 680 to

[2]K. P. Kolchin and E. L. Assur, "Graphitization of Strip Steel," *Journal*, Iron and Steel Institute. Abstract (British). Vol. 137, 1938, No. 1, p. 69A.

750 degrees Cent. (1255-1380 degrees Fahr.) but the abnormal steel graphitized readily even at so low a temperature as 600 degrees Cent. (1110 degrees Fahr.). This observation is in agreement with the records of the present investigation. The reason Kolchin and Assur, according to their report, found a limited amount of graphitization in the normal steel, can be explained on the basis that the atmosphere during heat treatment permitted the infiltration of some oxygen into the steel. The tests of the present authors[1] showed that the presence of oxygen during tempering at subcritical temperatures will cause graphitization in practically all of the steels herein considered.

In order to extend further possible correlative studies, hardenability tests were included in the investigation. Since hardenability is so largely dependent on grain size it was necessary to eliminate this factor in obtaining data for examination in relation to carbide stability or graphitizing tendency of the steels.

From the results of the two series of tests it is noted that the deeper hardening steels, H, L, J, R, and K, are higher in silicon plus manganese than any of the other alloys, and that these particular alloys have stable cementite. There is little reason to consider that the above mentioned elements have any effect in promoting the carbide stability, but their probable role in promoting deeper hardening introduces a factor which is not present in the other alloys. These other steels show hardenability values ranging from 0.17 inches down to 0.14 inches without discrimination between those having stable or graphitizing carbides.

DISCUSSION

Written Discussion: By Carl L. Shapiro, technical director, Lynnwood Laboratories, Inc., Newark, N. J.

The fine technique and the resulting conclusions of the authors are praiseworthy. In the writer's opinion, however, the authors were handicapped by lack of laboratory equipment and facilities, which resulted in using steel obtained from various outside sources and not made in the laboratory. Consequently, the original processing of these steels was unknown. Many factors, which influence the stability of cementite, abnormality, grain size and hardening, such as melting procedure, deoxidation practice, casting time and temperatures, cogging, rolling, soaking, in fact all the variables from selecting the furnace scrap to finishing the material, were perforce omitted. Indications are that the authors themselves are aware of this limitation since they say on page 349 that "No information could be obtained from the manufacturers

[1] loc. cit.

on the deoxidation practice——". It is to be regretted that as a result of inadequate laboratory facilities many fine technical investigations such as this one should suffer.

In the investigation the authors determine the relative degree of graphitization indirectly by the hardness of the steel. This method of determining graphitization is admittedly only relative in the same steel when given the same prior heat treatment and approximate in steels of the same chemical composition. However, the writer wonders why the amount of graphitic carbon was not determined chemically.

It is to be hoped that further investigation will be continued on these same steels at and above 950 degrees Cent. for various lengths of time in order to note:

(1) the effect of T^{max},

(2) the rate of grain growth in the coarsening temperature,

(3) the effect of T^{max} upon graphitization, and

(4) the effect of deoxidation upon grain growth.

Written Discussion: By H. A. Schwartz, manager of research, National Malleable and Steel Castings Co., Cleveland.

The authors' contribution is of interest both because of its theoretical implications and because of the attention currently being given to the graphitizable steels.

That aluminum promotes graphitization is, of course, well known but the authors have been able to draw a distinction between the effect of metallic aluminum and of the Al_2O_3 resulting from reactions within the metal. This latter furnishes a very welcome bit of information in a field which is as yet not well worked, namely whether aluminum acts, per se, as a remover of oxygen or by the formation of Al_2O_3 which by nucleation is the actual active agent. It is apparent in the present paper that within the authors' field of investigation the latter mechanism is the more important.

That the authors were unable to make satisfactory further correlations of graphitizing rate with other variables should not be too much of a cause for discouragement. If one variable predominates, it is rather too much to expect that statistical methods applied to so limited an amount of material should have permitted conclusions as to the effect of variables which are apparently of secondary importance. This is not to say that the authors' treatment of their data by statistical mathematics is not desirable for it furnishes the only means by which it could become known whether further conclusions might or might not be justifiable.

Oral Discussion

J. T. MacKenzie, Jr.:[3] In line with Dr. Schwartz' statement which we have just heard in regard to the secondary variables in the number of steels that were tested, it seems to me that something of this order might be cleared up by a few simple experiments.

Conclusion 2 in the paper reads: "In general, steels to which aluminum

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[3]United States Steel Corporation Research Laboratory, Kearny, N. J.

additions have been made and which show a fine-grained abnormal case in the McQuaid-Ehn test, will graphitize."

It would be interesting to know whether these same fine-grained steels would graphitize, and to what extent, if they were purposely made coarse-grained before attempting to graphitize.

In the same manner, conclusion 3: "With two exceptions, steels to which aluminum has been added but which remain coarse (normal) or even mixed grained (mixed normality) in the McQuaid-Ehn test do not graphitize."

If these coarse-grained steels were quenched from just above the Acm and thus the grain kept as fine as possible, would they still resist graphitization completely?

By these two series of experiments, the factor of grain size might be eliminated and in this manner the number of variables be reduced.

Authors' Reply

The authors wish to thank Messrs. Schwartz, MacKenzie, Jr. and Shapiro for their comments and suggestions.

Dr. Schwartz has raised the question as to whether aluminum, per se, or Al_2O_3 is the active agent in subcritical graphitization. In the case of steels where aluminum is added merely for deoxidation or grain control, it is believed that the deoxidation product Al_2O_3 is responsible, and that aluminum, per se, causes graphitization only when added in sufficient quantity that the resulting steel can be classified as an aluminum alloy steel. It is not considered that the removal of oxygen by aluminum is a factor, because our tests have shown that the presence of oxygen is favorable to graphitization below the A_1. When annealing at 670 degrees Cent. (1240 degrees Fahr.) is carried out in uncovered lead practically all 18 steels graphitize to some extent whereas in covered lead only 8 of the 18 steels show any signs of graphite formation.

Mr. MacKenzie, Jr. has raised a point which perhaps needs some clarification. It may appear contradictory to state that, in general, inherently fine-grained steels graphitize as compared to inherently coarse ones, and yet to affirm that fine grain steels graphitize faster the higher the temperature (and hence the larger the grain size) from which they are quenched prior to annealing at 670 degrees Cent. (1240 degrees Fahr.) This apparent anomaly is explained as follows: Whether or not any steel has a tendency to graphitize depends upon the formation of Al_2O_3 in suitable form, resulting from the deoxidation practice used in making the steel. Such practice, incidentally, also results in the steels which are generally fine-grained. Thus it is not the fine grain size but the mode of deoxidation which is responsible for graphitization. Since the temperature of quench has apparently no effect on the results of deoxidation practice, it might be assumed that modification of grain size as produced by heat treatment would have little effect on subsequent graphitization at subcritical temperatures. Experimental study has revealed that the higher the temperature of quench the more readily the steel will graphitize. The authors have evidence which strongly suggests that this effect, however, is not due to the progressive increase in grain

size but to the greater number of cracks produced in the steel as the quenching temperature is raised.

It is to be emphasized that the presence of Al_2O_3 in suitable form is the prime requisite for subcritical graphitization and that the cracks are a contributory factor. The authors have, as Mr. MacKenzie suggested, heated inherently fine-grained steels until they became coarse before quenching. Some increase in amount of graphitization did occur. Similarly the inherently coarse-grained steels on quenching after a short time from just above the Ac_1 failed to graphitize when subjected to prolonged annealing at 670 degrees Cent. (1240 degrees Fahr.).

In answer to Dr. Shapiro's intimation that the authors were handicapped by lack of laboratory equipment and facilities, it should be emphasized that such was not the case, for the department is unusually well equipped for all types of metallurgical research. Indeed, experimental melts have been prepared in the laboratory, by use of an Ajax Northrup high frequency furnace, in order to permit a check on the conclusions obtained from a study of the commercial steels. The commercial steels were all carefully selected by the various manufacturers and it was due to a misunderstanding alone that one of the companies did not have available the data desired on deoxidation practice. Thus Dr. Shapiro's partial quotation from the text of the paper is rather misleading, since such data are recorded on all but three steels. He is quite right, however, when he states that all of the variables of steel making practice were not accounted for.

No doubt analyses for graphite would be valuable but hardness tests, correlated with microstructural studies, appeared adequate for the conclusions drawn. A fairly close estimate of the degree of graphitization in a steel can be obtained by observing the quantity of undissociated cementite.

CORRELATION OF HIGH TEMPERATURE CREEP AND RUPTURE TEST RESULTS

By R. H. Thielemann

Abstract

It is desirable in the design of high temperature equipment to have a working picture of both the creep and the rupture properties of a material at the temperature in question before arriving at safe working stresses. Depending upon the particular application either creep or rupture after long periods of time may be the criterion of successful operation. With materials which fail by intergranular cracking after long periods of time, it is found that the amount of elongation preceding fracture may be greatly reduced. In such cases, the relation of the long-time rupture strength to the long-time creep strength of the material assumes added importance.

It is the object of this paper to present certain results from creep-rupture tests, which, it is believed, illustrate the relation between the creep and the rupture properties of certain types of alloys. The four materials for which results are reported are commonly employed in various types of high temperature equipment. The combined creep and rupture plots which are presented appear to have much merit in guiding the selection of stresses for high temperature engineering design.

FOR the determination of the long-time tensile properties of materials at elevated temperatures, the familiar creep and rupture tests are employed. Both types of tests yield information which is necessary for the judicious design of high temperature equipment. Where service requires that plastic deformation over long periods of time be limited to very small values, as in steam turbines and high temperature machinery, reliable creep test information on materials is of prime importance. In other types of equipment where the maintenance of such close dimensions is not necessary, larger values of deformation with time can usually be tolerated providing, of course, that failure does not occur. For such applications, the long-time rupture properties of materials should be taken into consideration.

Recent results have indicated that, with certain materials, the long-time load carrying ability may be influenced by the occurrence of intergranular cracking. With the occurrence of this phenomenon, the normal amount of deformation preceding fracture may be greatly reduced. The most brittle fractures seem to result from low stressed tests of comparatively long duration. Indications from these tests are that breaks will occur with even smaller values of elongation in tests of still longer duration. Robinson (1)[1] has discussed the significance of this trend with reference to design stresses and concluded that where the amount of creep is not in itself an important consideration, there is good reason to be guided by the rupture results rather than by the creep results. It appears that for all high temperature applications both the creep and rupture properties of materials should be known.

During the past 15 years a large amount of effort has been expended on creep testing, and much of the available information has been published in the literature (2). Recently, results from high temperature rupture tests on various types of materials have also been made available (3), (4), (5), (6), but there has been no correlation between these results and the creep results that could be used for the judicious selection of safe working stresses. It is the object of this paper to present certain results from creep-rupture tests, which, it is believed, illustrate the relation between the creep and rupture properties of certain types of materials. Combined creep and rupture plots are presented which appear to have much merit in guiding the selection of stresses for high temperature engineering design.

Apparatus and Technique

A complete description of the equipment employed in the Research Laboratory of the General Electric Company for making high temperature-sustained load rupture tests appeared in previous papers (5), (6) and need not be repeated.

At the present time, seven of the twelve bar machines are in operation and tests are being conducted at 900, 1000, 1100, 1200 and 1300 degrees Fahr. (480, 540, 595, 650, 705 degrees Cent.). Although various shapes and sizes of test specimens have been used, two sizes have recently been standardized, each employing a 4-inch gage

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[1] The figures appearing in parentheses refer to the bibliography appended to this paper.

length; one 0.253 inch in diameter and the other 0.357 inch in diameter. These threaded specimens are machined from bar stock, the larger diameter specimens being employed at the higher temperatures for materials that have relatively poor resistance to oxidation. With either size of test specimen, results are very consistent and where oxidation is not a factor, good agreement is found from tests of the same material using both diameters.
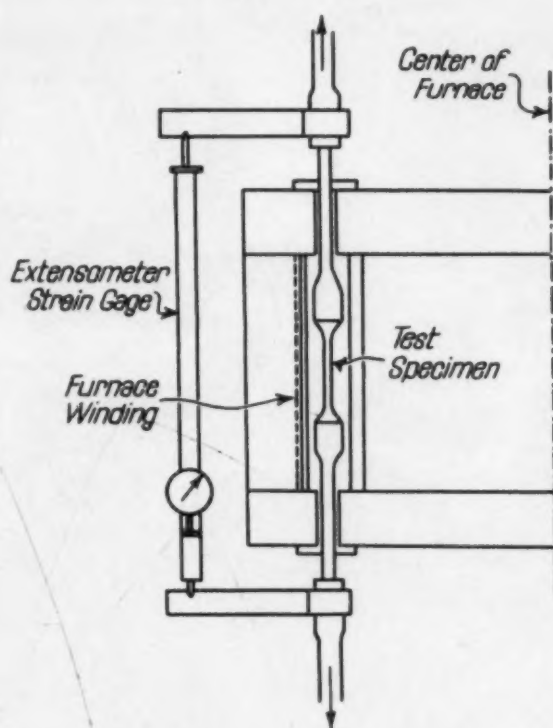
Fig. 1—Cross Section of Furnace Showing Location of Test Specimen and Method for Making Creep Measurements.

Briefly, the testing procedure is to prepare at least six test bars of the material for the tests at each temperature. These bars are placed in the rupture furnaces and different values of stress are applied to each bar. Creep measurements are taken during the progress of the test for the long-time tests. The tests are run until failure has occurred, and the time required for fracture is thereby obtained.

Fig. 1 shows diagrammatically the location of the test specimen in the cylindrical rupture furnace and indicates the method that is employed for making the creep measurements. The extensometer strain gage is shown in position at one station of the 1300 degrees Fahr. (705 degrees Cent.) furnace in Fig. 2. It should be men-

tioned that in making creep measurements in this manner, the appa-
ratus is such that any elongation that occurs in either the threads
or the test bar holders is included in the overall reading. But due to
the construction of the stainless steel holders and the machines, this
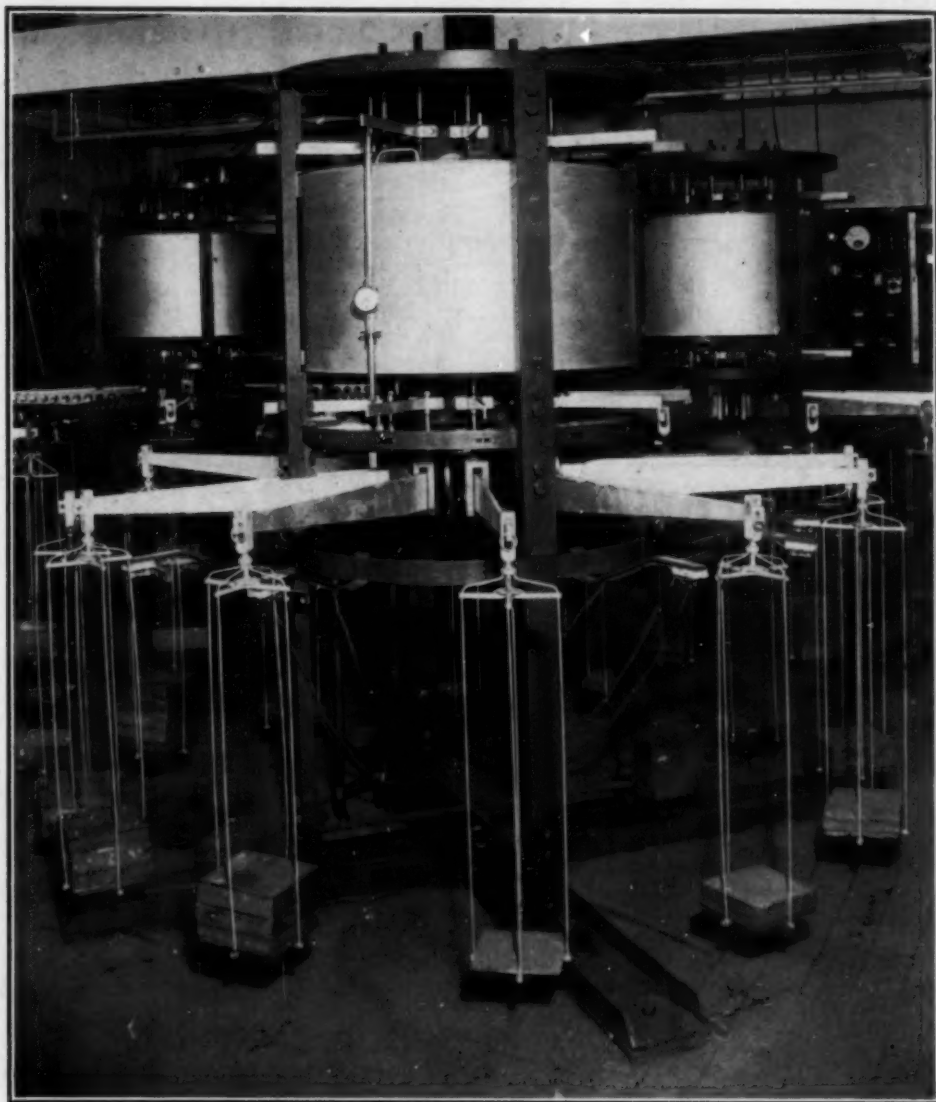
Fig. 2—Rupture Testing Equipment with Extensometer Strain Gage in Place
for Making Creep Measurements.

error would have to be very small. Also, the error would be additive,
so that the material would appear to be weaker in creep than it really
is and the reported creep values would be conservative. The method
has the advantage in that it is simple and affords an accurate means
of making several readings in a comparatively short period of time.

Discussion of Results

A combined plot showing both creep and rupture test results at 900 degrees Fahr. (480 degrees Cent.) for a nickel-chromium-molybdenum steel is shown in Fig. 3. Both ordinates of the plot are
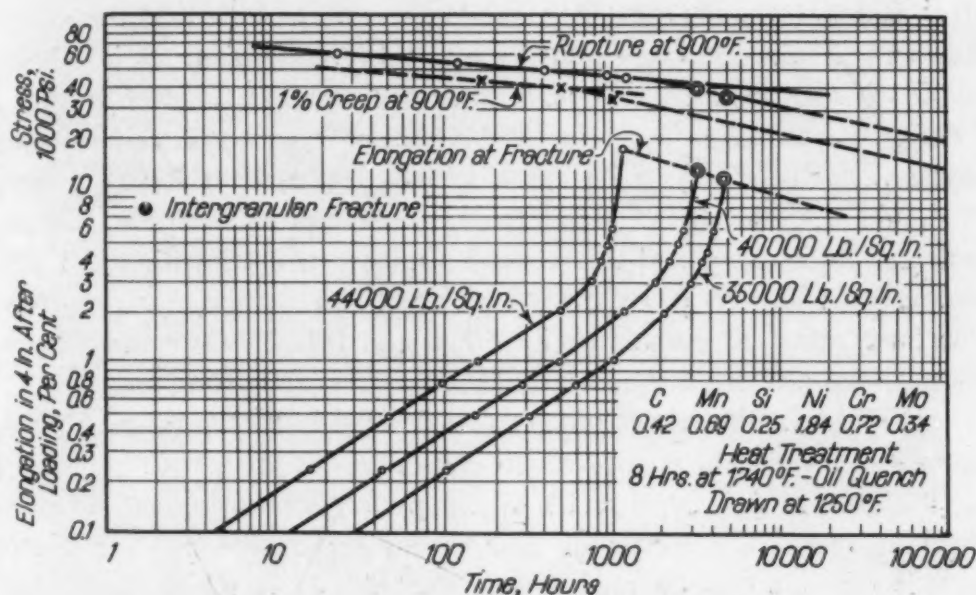
Fig. 3—Creep Rupture Results at 900 Degrees Fahr. for Nickel-Chromium-Molybdenum Steel.

logarithmic. The rupture curve at the top is obtained by plotting the stress against the time required for failure to occur. It will be noticed that for all tests up to about 2000 hours of duration, ductile transcrystalline type fractures resulted. Fractures for the tests of longer duration were of the intergranular type. The creep curves at the bottom of the sheet were obtained by plotting the elongation after loading, in per cent, against the time, in hours. Up to about 2 per cent of plastic elongation, these curves appear to be almost straight lines. It is interesting to note that with the longer time tests, the elongations at fracture appear to be successively less. A rather uncertain extrapolation of the elongation at fracture line indicates that fracture in 100,000 hours would occur with about 5 per cent elongation in a 4-inch gage length. The 1 per cent creep curve, which is plotted beneath the rupture curve at the top, is obtained by plotting the stress against the time required for 1 per cent of plastic elongation. The latter values are obtained from the creep curves at the bottom of the plot. Thus, the two curves are different

in that the rupture curve indicates the time that is required, with a given stress, for failure to occur whereas the creep curve indicates the time, with the same stress, that is required for 1 per cent of creep.

By extrapolation, the indicated stress that would produce failure after 100,000 hours is 20,000 pounds per square inch and the stress that would produce 1 per cent of creep in 100,000 hours is 13,500 pounds per square inch. Thus, with this material at this temperature, a design stress based upon the 100,000 hour 1 per cent value would have a factor of safety of 20,000/13,500 or 1.48 against actual failure. It should be pointed out that the 1 per cent creep curve as shown indicates total creep and this should not be confused with the 1 per cent creep rate per 100,000 hour value that is commonly reported. The latter may be greater or less than the former, depending upon the material and the temperature.

While this plot leaves much to be desired in the manner of completeness, it presents results which can be easily applied to engineering design. If desired, a series of creep curves showing the stresses required for either larger or smaller values of creep after given periods of time can be readily constructed. It is unfortunate that results from tests of still longer duration are not available, so that the extrapolation of the 1 per cent creep curve could be accomplished with somewhat greater certainty. This, however, serves to emphasize the desirability for long-time tests.

The original heat treated structure of this material before test is shown in Fig. 4. The appearance of the fracture for the test of longest duration, as shown in Fig. 5, indicates an absence of hot ductility.

A similar plot showing rupture and creep results at 1000 degrees Fahr. (540 degrees Cent.) for a heat treated chromium-molybdenum-vanadium alloy is shown in Fig. 6. This is an example of a material that has a high creep strength and, in addition, is at this temperature notably susceptible to failure by intergranular cracking. The extrapolated rupture and 1 per cent creep curves, shown at the top of the chart, indicate that the 100,000 hour 1 per cent creep and rupture strengths are identical. It follows that the elongation at fracture for the 100,000 rupture point should be in the neighborhood of 1 per cent. That this would be the case is also indicated by trend of the elongation at fracture line. In comparing the creep curves for this material with the creep curves shown in Fig. 3, one factor stands
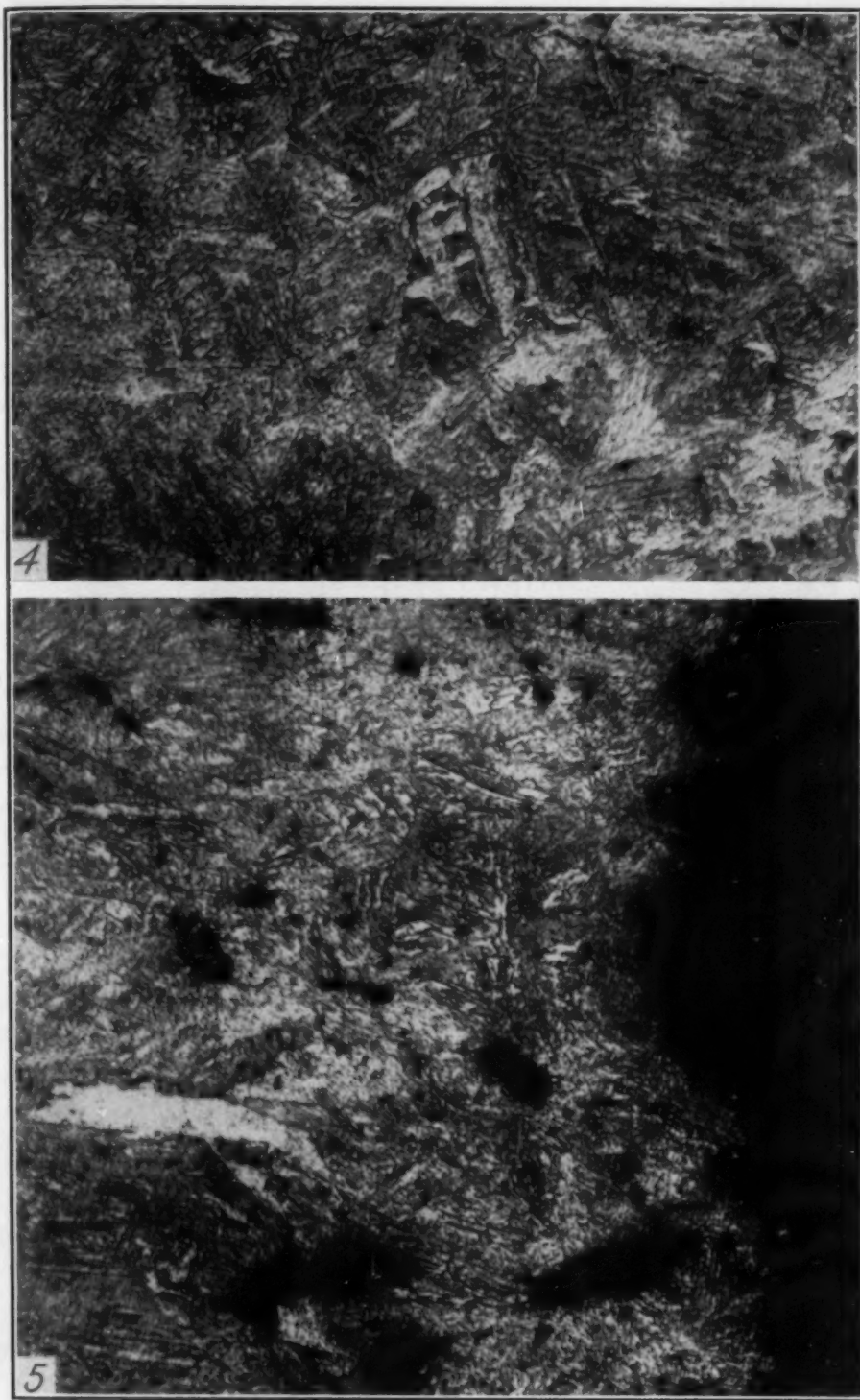
Fig. 4—Structure of Nickel-Chromium-Molybdenum Steel Before Testing. Nital Etch. × 250.

Fig. 5 — Intergranular Fracture of Nickel-Chromium-Molybdenum Bar. Stressed at 35,000 Pounds Per Square Inch for 4990 Hours at 900 Degrees Fahr. Elongation —12.5 Per Cent in 4 Inches. Reduction of Area 37 Per Cent. Nital Etch. × 250.

out which is noteworthy. Whereas, with the nickel-chromium-molybdenum material, the creep curves were fairly straight up to about 2 per cent elongation, the curves for the chromium-molybdenum-vanadium material indicate a marked change in creep rate at
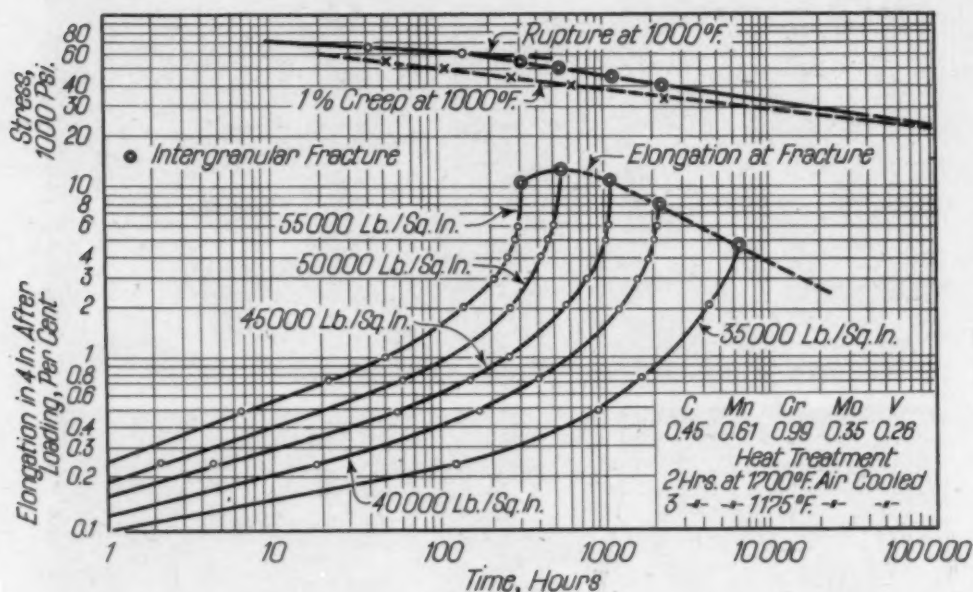
Fig. 6—Creep Rupture Results at 1000 Degrees Fahr. for Chromium-Molybdenum-Vanadium Steel.

elongations of less than 1 per cent. Whether or not this change of slope indicates the occurrence of incipient intergranular cracking is not known, but work is now being started to determine whether or not this is the case. It should be noted, however, that for the latter material, both the 100,000 hour creep strength and the 100,000 hour rupture strength are higher at 1000 degrees Fahr. (540 degrees Cent.) than are the corresponding creep and rupture stengths of the former material at 900 degrees Fahr. (480 degrees Cent.).

The original heat treated structure of this material before test is shown in Fig. 7 and the appearance of the longest time fracture is shown in Fig. 8.

Similar creep and rupture results for a carbon-molybdenum steel at 1000 degrees Fahr. (540 degrees Cent.) are shown in Fig. 9. Here again, the factor of safety against actual failure based on
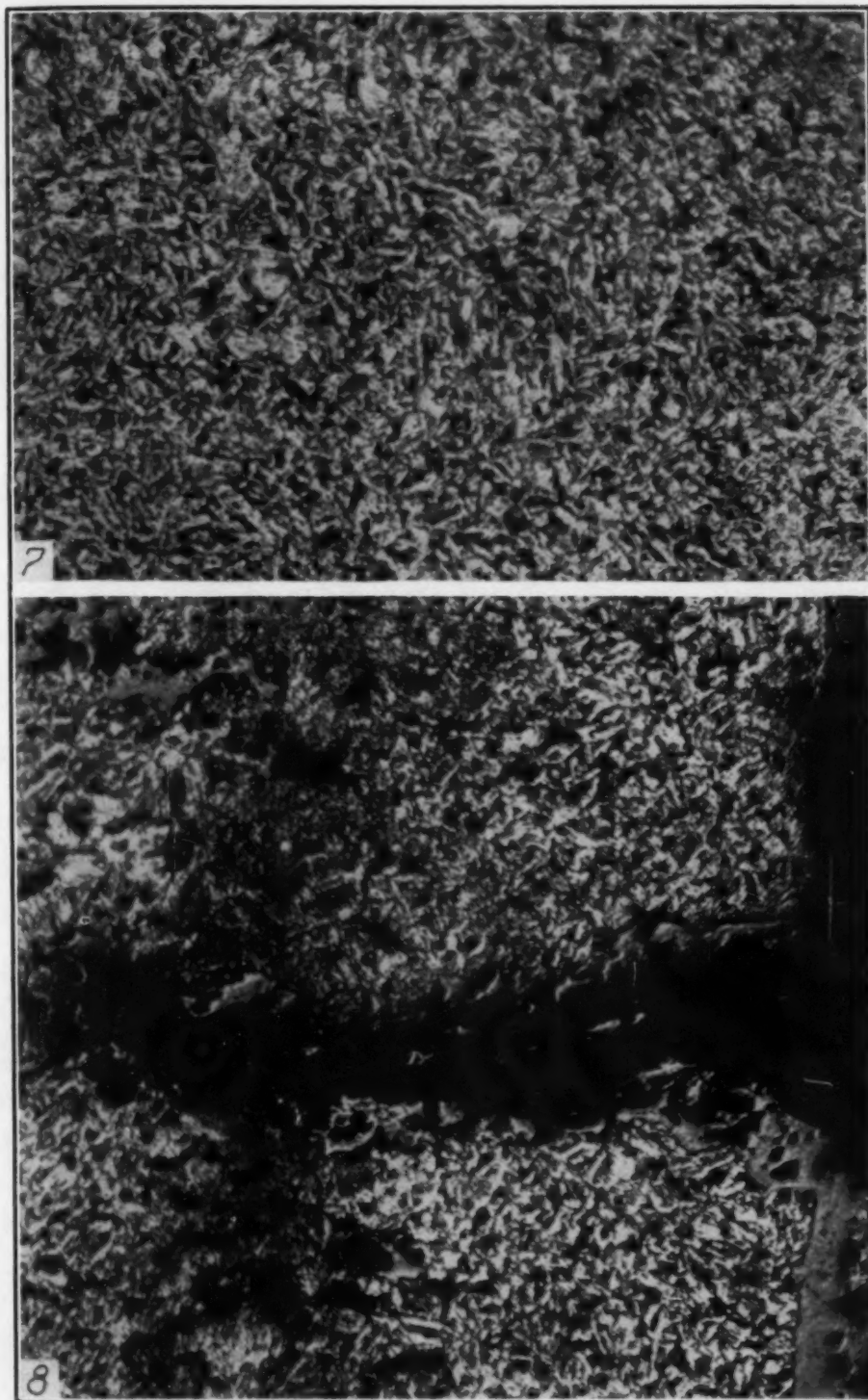
Fig. 7—Structure of Chromium-Molybdenum-Vanadium Steel Before Testing. Nital Etch. ✕ 250.

Fig. 8—Section Near Fracture of Chromium-Molybdenum-Vanadium Bar. Stressed at 40,000 Pounds Per Square Inch for 2330 Hours at 1000 Degrees Fahr. Elongation 8 Per Cent in 4 Inches. Reduction of Area 15 Per Cent. Nital Etch. ✕ 250.

the stress that will produce 1 per cent of creep in 100,000 hours is comparatively low.

For this material, the ASME boiler code (Specification S45-P1 (7) allows a working stress of 5000 pounds per square inch for steam pipes. On this basis, the 100,000 hour rupture strength of 9000
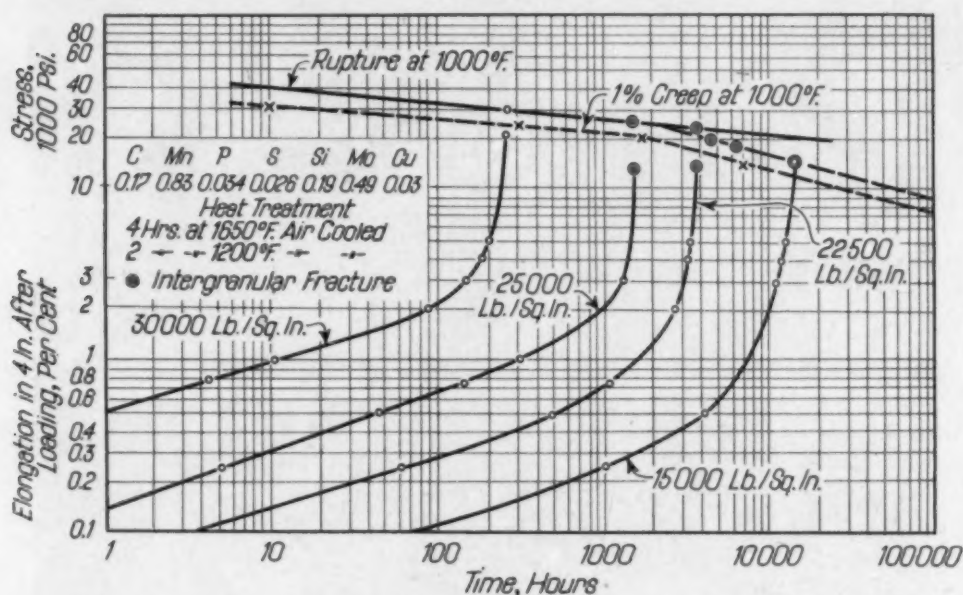
Fig. 9—Creep Rupture Results at 1000 Degrees Fahr. for Carbon-0.5 Per Cent Molybdenum Steel.

pounds per square inch gives a factor of safety against failure slightly less than 2.

For this particular composition, the data do not preclude that the elongation at fracture for the 100,000 hour point would be much less than 10 per cent in a 4-inch gage length. Rupture tests on similar compositions have indicated much lower values (1). It is felt that the comparatively high values of elongation accompanying the longtime fractures with this particular composition might be due to the increased manganese content which the other materials did not have. The original heat treated structure before testing is shown in Fig. 10, and the appearance of the longest time fracture in Fig. 11. Although this fracture was mostly of the intergranular type, it is evident from the elongated grains that ductility is not completely lacking.

For the purpose of comparison, creep and rupture results for a material which is apparently not susceptible to failure by intergranular cracking at 1200 degrees Fahr. (650 degrees Cent.) are shown in Fig. 12. Although these tests are of shorter duration and still in-

Fig. 10—Structure of Carbon-0.5 Per Cent Molybdenum Steel Before Test-
ing. Nital Etch.   × 250.
Fig. 11—Structure at Fracture of Carbon-0.5 Per Cent Molybdenum Steel.
Stressed at 17,500 Pounds Per Square Inch for 6091 Hours at 1000 Degrees
Fahr. Elongation—17.4 Per Cent in 4 Inches. Reduction of Area 46 Per Cent.
Nital Etch.   × 250.

Table I
Rupture and Creep Properties of Materials Tested

Material	Temperature of Test °F.	(S_R) 100,000 Hour Rupture Stress Lb. Per Sq. Inch	(S_C)* 100,000 Hour 1% Creep Stress Lb. Per Sq. Inch	Ratio S_R/S_C
Ni-Cr-Mo (Fig. 3)	900	20,000	13,500	1.48
Cr-Mo-V (Fig. 6)	1000	22,000	22,000	1
C-½% Mo (Fig. 9)	1000	9,000	7,300	1.23
7% Cr-Mo-Si (Fig. 12)	1200	3,750	1,700	2.20

*Indicates total creep and should not be confused with 1 per cent creep rate per 100,000 hour value that is commonly reported.

complete, there is no reason to believe that long-time fractures will occur with less than 30 per cent elongation. In addition, the factor of safety against failure based on the stress that will produce 1 per cent of total creep in 100,000 hours is greater than 2. It is interesting to note that the 1 per cent creep curve parallels the rupture curve.
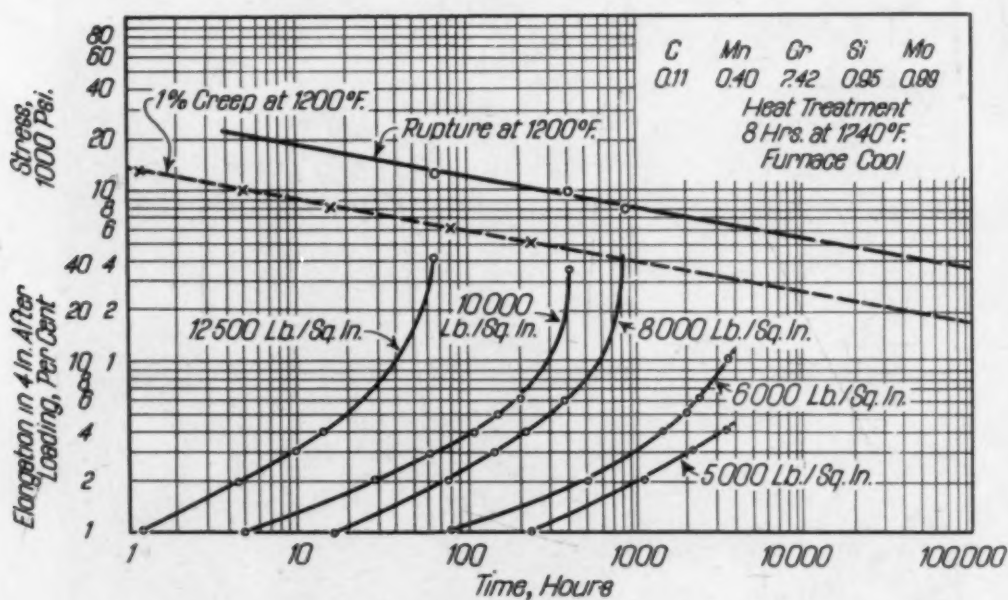
Fig. 12—Creep Rupture Results at 1200 Degrees Fahr. for 7 Per Cent Chromium Steel.

This same behavior is found for other materials which do not fail by intergranular cracking in tests of long duration. Photomicrographs showing the original structure and the appearance of the longest time fracture are shown respectively in Figs. 13 and 14. With the transcrystalline type of fracture the break occurs through the
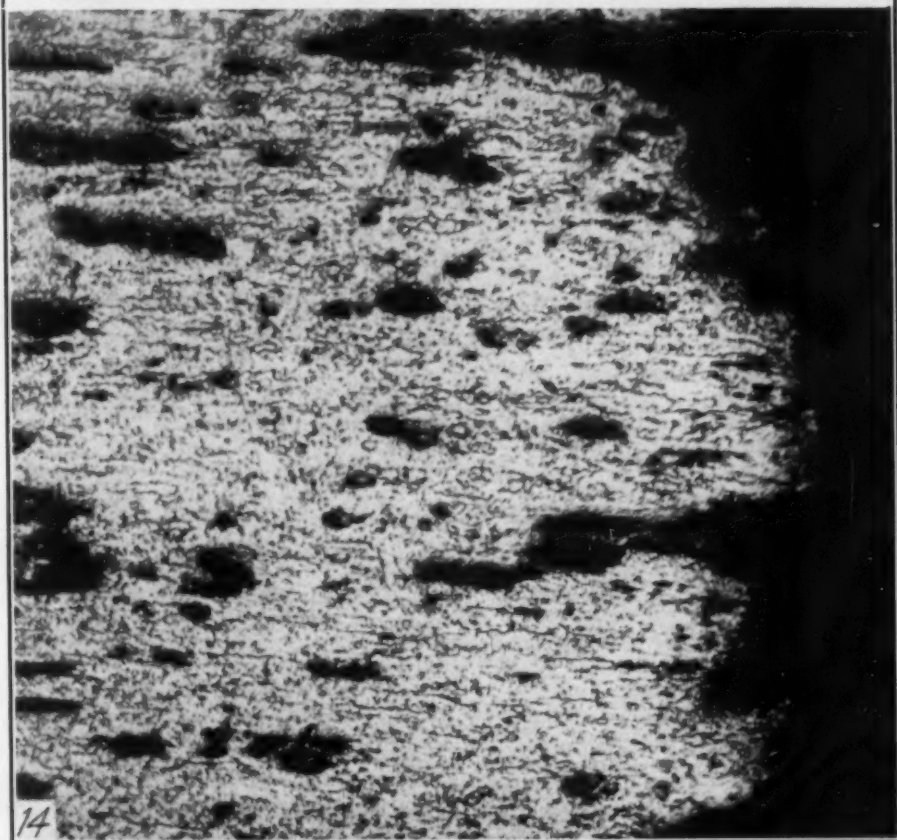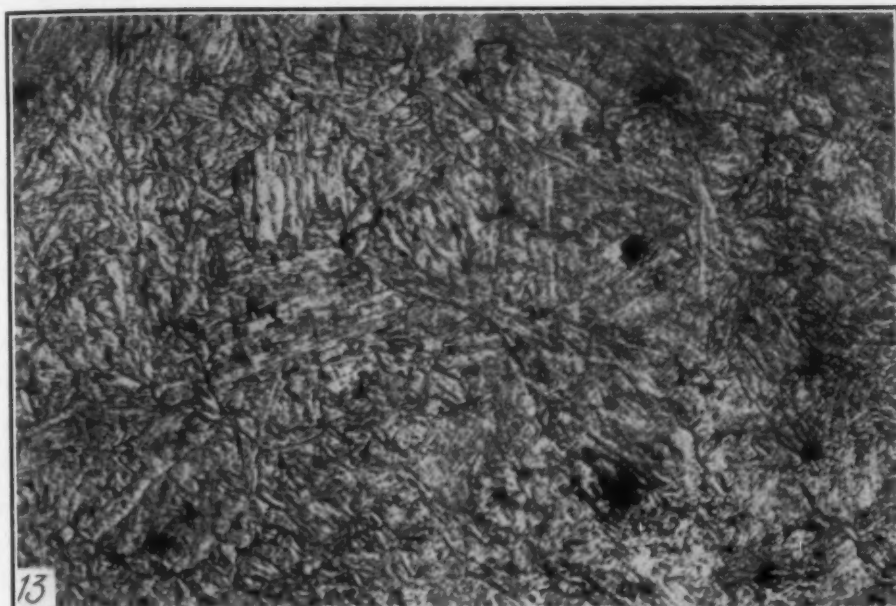
Fig. 13—Structure of 7 Per Cent Chromium Steel Before Testing. Nital Etch. × 250.

Fig. 14—Structure at Fracture of 7 Per Cent Chromium Steel. Stressed at 8,000 Pounds Per Square Inch for 806 Hours at 1200 Degrees Fahr. Elongation in 4 Inches 40 Per Cent. Reduction of Area 84 Per Cent. Nital Etch. × 250.

grains, and for ductile materials the grains are elongated in the direction of the load as shown in the latter figure.

The composition and heat treatment for each of the materials tested are indicated in the respective figures. Results from the extrapolated creep and rupture curves for each material are shown in Table I.

CONCLUSIONS

From these results it appears that both the creep and rupture properties of a material must be known before the behavior over long periods of time at elevated temperatures can be predicted with any degree of certainty. A combined plot of creep and rupture test results, as presented, is found to be useful in correlating the creep and rupture properties. This method of plotting affords a convincing means of arriving at safe working stresses for design purposes.

With some materials there is good reason to believe that the amount of elongation accompanying long time fractures will be very small indeed. Usually, these materials possess high creep strengths and fail by a process of intergranular cracking. As indicated by the results, for materials of this type, long-time factors of safety against failure may be almost lacking if design stresses are based on the value that will produce 1 per cent of creep in 100,000 hours. On the same basis, materials that do not fail by intergranular cracking in tests of long duration usually show much larger factors of safety against failure.

APPRECIATION

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3. A. E. White, C. L. Clark, and R. L. Wilson, "The Fracture of Carbon Steels at Elevated Temperatures," TRANSACTIONS, American Society for Metals, Vol. XXV, Sept. 1937, p. 863-888.

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5. R. H. Thielemann and E. R. Parker, "Fracture of Steels at Elevated Temperatures After Prolonged Loading," *Transactions,* American Institute of Mining and Metallurgical Engineers, Vol. 135, 1939, p. 559-582.

6. R. H. Thielemann, "Some Effects of Composition and Heat Treatment on the High Temperature Rupture Properties of Ferrous Alloys," Preprint No. 46 for 1940 Annual Meeting, American Society for Testing Materials.

7. ASME Boiler Code, Revisions and Addenda. *Mechanical Engineering,* June 1940, p. 486-487.

DISCUSSION

Written Discussion: By Parker P. King, metallurgical department, Bethlehem Steel Co., Bethlehem, Pa.

Mr. Thielemann is to be congratulated on an excellent presentation of some very interesting high temperature data. The rupture curves plotted on the same logarithmic paper as the creep curves is a novel method of presenting the combined data which appears to have considerable significance and possibly a great deal of practical value.

For instance, in Fig. 6, creep-rupture results on a bolting steel (ASTM-A193-39T-B14 type) are shown. The data indicate that this steel, heat treated by normalizing and drawing, would fail after approximately ten years of service at 1000 degrees Fahr. if loaded at a stress of 23,000 pounds per square inch. In the meantime, this stress would have produced only 1 per cent elongation in a 4-inch gage length over the 10-year period. In actual service, it is doubtful whether such an impending failure could be detected beforehand, since 1 per cent elongation is probably insufficient to cause a leak in a bolted flange connection installed in a steam line over a 10-year period.

Thus we may have a dangerous condition of embrittlement taking place in high temperature service in such steels. It will indeed be most interesting to obtain this same type of data on various other bolting steels whose rupture properties have not as yet been reported by Mr. Thielemann or others. There are some old and tried grades of bolting steel which to date have never failed in the field of high temperature service. If rupture tests indicate that these steels are not subject to relatively brittle failure with low elongation, the tests will have even increased significance. However, if the data should show the reverse to be true, we might feel that conditions are developed in the rupture test which are not encountered in service.

It is possible by the method of plotting the test data used in this paper to determine in the same chart not only the stress to cause 1 per cent total creep in 100,000 hours, but also the stress to cause rupture in 100,000 hours as well as the amount of elongation at rupture. The relationship among the three variables seems to be rather well defined and there may be a tendency to assign exact values to a grade of steel on the basis of one complete series

of tests. Having in mind the history of creep testing in which undue accuracy was ascribed to early results and not forgetting that the present method involves considerable extrapolation, we should be inclined to regard the data as approximate until they have been well verified.

It is my opinion that high temperature tests, no matter what variety, are of academic interest only unless they tie in directly with service performance in the field. The art of high temperature testing has developed at a fast rate over the past ten years, accompanied by constant changes in technique and test methods. High temperature test equipment seems to become obsolete as rapidly as last year's automobile. We are still searching for a short-time test from which we may forecast the performance of a steel during a service life at high temperature. Mr. Thielemann's paper appears very promising in this direction and with further investigational work along the same lines it may represent a very valuable contribution to the problem of how to test a high temperature steel.

Written Discussion: By R. G. Sturm, research engineer physicist, Aluminum Company of America, New Kensington, Pa.

This brief paper presents creep data of steel at high temperatures in a very effective way. The use of total creep in the discussion of creep values avoids the difficulties encountered in quoting creep rates. The author's data confirm the dependence of the creep rate upon the elapsed time at which the creep rate is determined. Creep tests[1] on a number of materials at room temperatures including steel, copper and aluminum alloys show that the plot of log-creep against log-time has the same general characteristics as the curves shown by the author, i.e., an essentially straight line with a slope less than unity until failure impends and then a curved portion turning upward. The earlier paper showed one exception to this rule at room temperatures. This exception is hard drawn galvanized steel wire which did not show a tendency to curve upward when failure was impending.

It is indeed interesting to note that for creep values below those occurring at impending failure, the slope of the curves shown by the author is practically the same for all stresses on a given material at any given temperature. The difference in characteristics of the curve for galvanized steel wire at room temperatures given in the earlier paper and the curves obtained for the steels by the author might well lie in the fact that the rate of strain hardening[2] of steel at room temperatures is greater than unity whereas the rate of strain hardening at the elevated temperatures is probably less. The fact that the rate of strain hardening for aluminum alloys is also less than unity together with the similarity in creep characteristics between aluminum alloys at room temperatures and steel at higher temperatures, suggests that the different metals at the same homologous temperature (percentage of the melting point in absolute scale) exhibit the same general characteristics in response to stress.

The significance of the author's use of total creep values instead of creep

[1]R. G. Sturm, C. Dumont and F. M. Howell, "A Method of Analyzing Creep Data," *Journal of Applied Mechanics,* June 1936, p. A-62.

[2]R. L. Templin, "Effect of Cold Working on Physical Properties of Metals," American Institute of Mining and Metallurgical Engineers Technical Publication No. 238, Class E, Institute of Metals, No. 84, September, 1929.

rate is quite apparent when considering deformation at rupture. The significance of total creep in controlling design is often lost in what the writer believes is an over-emphasis on creep rate. Engineering structures become unusable from excess deformation more often than from an excess of rate of deformation. For this reason the use of total creep or factors indicative of total creep has real significance in engineering design.

Oral Discussion

A. V. DEFOREST:[3] It seems to me that this intergranular cracking as a function of high-temperature service is one that is extremely important and has been overlooked by laboratories for a considerable length of time. Those dealing with failures in the field found long ago that the important failures were due to intercrystalline cracking and not to a direct failure of the material in creep as such.

It would appear that that type of failure is very closely related to the corrosion stress problem, in which the presence of stress changes the type of corrosion from general to intergranular, frequently with disastrous results.

The evaluation of the strength of metals under combined temperature and stress plus corrosion from the surrounding atmosphere adds greatly to the complexity of the creep picture as a whole. It would seem that separate creep tests should be run using an inert gas for the temperature strain rate plot and an atmosphere of the same type as met in service for the study of intergranular corrosion effects. It is fortunate that Mr. Thielemann has drawn attention to the great importance of this last phase of the general problem.

Author's Reply

The comments presented by those who have discussed this paper are both timely and significant.

In replying to Mr. King's question of whether or not the high temperature rupture test results will accurately indicate the performance of materials in bolting service, certain factors should be given consideration. In many types of bolting practice, the residual operating stress in a bolt depends to a large extent on the relaxation properties of the material at the operating temperature. The high initial stress set up by tightening is usually greatly decreased even after comparatively short periods of time. If the residual or operating stress is less than the long-time constant stress rupture strength, it seems quite likely that failure would never occur. What the effect of subsequent tightenings might be on the long-time rupture properties of a bolt has not yet been determined. It seems likely, however, that the accumulated elongation resulting from relaxation after each tightening should result in higher elongation at fracture than would be obtained from a low stressed constant stress rupture test on the same material and at the same temperature.

In so far as the type of fracture to be encountered is concerned, there is every reason to believe that if a material breaks by intergranular cracking in

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[3]Professor of mechanical engineering, Massachusetts Institute of Technology, Cambridge, Mass.

the rupture tests it will fail by the same manner under comparable stress conditions in service. It follows that materials which are extremely susceptible to failure by intergranular cracking in long-time tests at a given temperature may fail without warning in bolting service at that temperature. The occurrence of intergranular cracks as at the roots of threads, would greatly decrease the safe long-time load carrying ability.

Mr. King's suggestion that short-time rupture tests might be employed as a rapid means of evaluating the high temperature properties of a material is indeed pertinent. There is good reason to believe that once the rupture properties of a given material have been established from long-time tests, short-time tests of perhaps a hundred hours duration on subsequent heats of the same material could be made for verification. The author feels, however, that there is no short-time test that will be satisfactory to evaluate the long-time properties of a new and previously untried material.

What Mr. deForest has said about the problem of intergranular cracking and the effect of intergranular corrosion on the load-carrying ability is indeed important. The author agrees that the effect of the surrounding atmospheres, as met in service, will add greatly to the problem of interpreting the long-time performance of materials which are susceptible to such. It would be highly desirable to duplicate, as near as possible, service conditions in laboratory tests.

Mr. Sturm's criticism of the over-emphasis placed on the evaluation of materials at elevated temperatures by creep rate determinations is welcome indeed. To quote creep rate values, as determined at a given testing time (usually 1000 hours) without regard for total elongation at that time is certainly misleading. Such values, if determined at a fixed amount of elongation rather than at a fixed period of time would be much more significant, but even these would not be a true indication of long-time performance. Depending upon the material and the temperature, the stress that will produce 1 per cent of total creep in a hundred thousand hours may be greater or less than the stress that will give a creep rate of 1 per cent per 100,000 hours at the 1000-hour period. Furthermore, since there is probably no such thing as a constant creep rate over long periods of time, creep rate values are without significance. As Mr. Sturm suggests, engineering structures become unusable from excess deformation more often than from an excess rate of deformation.

Mr. Sturm's references to room temperature creep results on galvanized wire, copper, and aluminum alloys are indeed interesting. The fact that an approximate straight line relation from a plot of log-creep against log-time is found for so many metals suggests strongly that the creep mechanism of all metals may be similar with response to stress. We have found the same straight line relation from creep tests on lead and other nonferrous alloys at room and elevated temperatures.

CHAFING FATIGUE STRENGTH OF SOME METALS AND ALLOYS

By G. Sachs and P. Stefan

Abstract

Endurance tests on cylindrical test bars offer a simple method for determining the effect of stress raisers on the fatigue strength. The resulting chafing fatigue strength appears closely related to the notch fatigue strength obtained on specimens with very severe notches. The effects of cold work and heat treatments on the regular and on the chafing fatigue strength respectively are very different. Generally, annealed wrought metals have a higher chafing fatigue strength but a lower regular fatigue strength than the harder cold-worked and heat treated conditions. Cast steels and aluminum alloys may have a higher chafing fatigue strength than wrought metals of the same type.

THE majority of service failures of moving machinery is caused by fatigue. According to earlier conceptions no fatigue failure of metals will be encountered if the stress cycles stay within certain limits, known as the fatigue strength. Numerous extensive investigations have therefore been carried out to determine the fatigue strength values of the commercially important metals and alloys under varying conditions.

However, the service conditions may differ considerably from those which must be applied in laboratory tests to obtain reproducible strength values. Thus, necked and polished specimens are generally used to determine the "regular" fatigue strength, but the values obtained in this manner have often been found to be much higher than those of specimens which closer approach the commercial conditions.

Several of the factors which may substantially reduce the fatigue strength are probably related to each other, the source of their harmful effect being local accumulations of stresses that are considerably higher than the conventionally determined maximum tension, compression, or shear stresses. There exist three, and probably only three, types of such "stress raisers":

1.   "Notches," such as grooves, holes, keyways, section changes, and also rough surface conditions, are "stress-raisers" proper because they create local stress concentrations at the surface.   Inherent notches, such as slag inclusions, blowholes and shrinkage cracks (flakes) act fundamentally in the same manner.

2.   Corrosion (simultaneously with fatigue) also decreases the fatigue strength according to general opinion because of the development of corrosion cracks, thus introducing stress raisers of the same type but even more severe than machined notches.
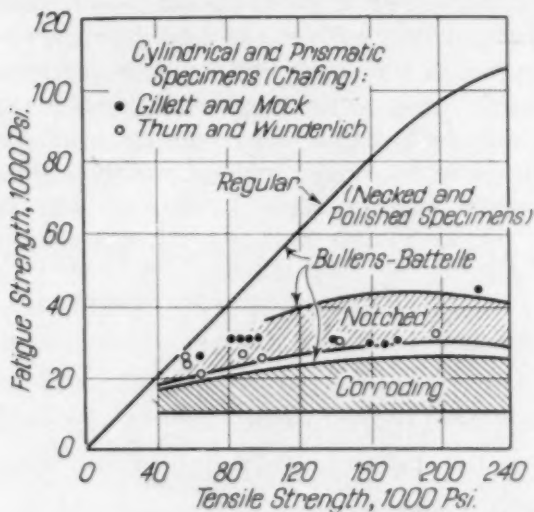
Fig. 1—The Effect of Different Stress Raisers on the Fatigue Strength of Various Steels.

3.   Lateral pressure, connected with a "chafing" or "fretting" on the surface of parts subjected to vibrating stresses, such as are present in press fits and bearing seats, exert also a severe reduction of the fatigue strength.

The close relation between the three types of stress raisers is not generally recognized as yet.   It is however well known (1)[1] that notches, as well as corrosion, reduce the conventional bending fatigue strength of steels the more the higher the tensile strength and the regular fatigue strength of the steel (Fig. 1).   It has been similarly observed by Gillett and Mack (2) in an early stage of the development of fatigue testing methods, that cylindrical and prismatical specimens break under chafing action within the grips and yield a low fatigue strength, which depends little upon the tensile and regular fatigue strengths of the steel (Fig. 1).   Similar results have been

[1]The figures appearing in parentheses refer to the bibliography appended to this paper.

obtained by Thum and Wunderlich (5). Thus, in any one of these three special types of fatigue little benefit is derived from alloying and heat treatment.

This fact is probably of a more general importance than has been recognized. Several instances have become rather famous in which the replacement of soft steels by heat treated high strength steels resulted in fatigue failures which have never been adequately explained. Applications in which alloying and heat treatment failed to improve fatigue strength in service are piston rods of diesel-driven ocean liners (3) and wire for suspension bridges (4). In the first case a combination of stress raisers, such as notching, corrosion and residual stress was found to exist, while in the latter case the presence of a chafing action may have contributed to the failure.

The analogy between the three types of stress-raisers also holds true for the beneficial effect of local plastic deformations at the inherent location of the fatigue fracture. Various processes such as surface rolling, peening and shot-blasting can be used to reduce the harmful effect of any of these stress-raisers.

While the effects of different notch shapes and corrosion agents on the fatigue strength have been extensively studied, only a few investigations have been carried out up to the present time on chafing fatigue. The fatigue strength of steels (2), (5), (6), aluminum alloys (7), and magnesium alloys (7), (8) is considerably reduced by any chafing action, such as are present in press fits, bearings, hubs or jaws. Surface rolling eliminates this reduction in the case of steels (5), (6) and magnesium alloys (8) to an extent which depends upon the rolling conditions and which may be increased by a fabric insert (8). Shotblasting also exerts a beneficial effect on the chafing fatigue strength of steels (9), (10). It has been furthermore shown by Sachs (8) that rotating bending tests with cylindrical specimens yield results corresponding to those obtained with other types of chafing and this fact renders the above discussed results of Gillett and Mack (2) on different steels particularly valuable.

The discussed investigations (2), (8) suggest that the testing of cylindrical (or prismatical) bars offers a simple procedure for determining chafing fatigue strength values. Besides, chafing fatigue appears to simulate a particularly sharp notch and thus to yield a desired limiting condition. If the relation between notch fatigue strength and chafing fatigue strength is to be more definitely established, such fatigue tests on cylindrical specimens may also replace

the fatigue tests of notched specimens which are difficult to produce uniformly.

The regular fatigue strength is usually improved by both cold work and heat treatments, which also increase the tensile strength of the metal. This is particularly true for the steels, the fatigue strength of which is for all conditions about 50 per cent of the tensile strength (Fig. 1). There is not much information available on the effect of different treatments on the fatigue strength of other alloy types; and there exists no simple relation between fatigue strength and tensile strength except for steels. The fatigue strength of cold-worked and heat treated metals in the presence of stress raisers has not created much attention as yet. However, Brophy (11) has shown that the notch fatigue strength of steels is affected by heat treatment in a very different manner from the regular fatigue strength.

The present investigation has therefore been carried out for the purpose of increasing our present knowledge on the behavior of cylindrical specimens in fatigue tests. It has been attempted in Part I of this paper to establish the effects of alloying, cold work and heat treatments on the chafing fatigue strength of wrought alloys. In Part II a few cast and wrought metals of similar compositions and conditions were compared.

Equipment

The machine used in this investigation was the well known Farmer rotating beam type of machine, producing a complete reversal of stress during each cycle and a uniform bending moment on the specimen. The speed of testing was approximately 3500 cycles per minute.

The common, slightly-tapered grips, made of cold drawn S.A.E. 1040 steel, were replaced in an early stage of the investigation by a special grip or collet (Fig. 2), made from oil-hardened S.A.E. 1040 steel. These grips were tightened on the specimen by drawing them in by means of a bolt threaded into the end of the grip and pulling against a corresponding shoulder in the sleeve. No effect was observed due to this change in the shape and hardness of the grips.

The grips gave a tight, chafing press fit, and with all the metals investigated some chafing and development of powdered reaction products was observed.

The grip pressure could not be varied intentionally but had to be always high enough to prevent any loosening of the specimen in the grips. There is some indication, however, that the pressure necessary to achieve this purpose must be rather high, as the observed chaf-
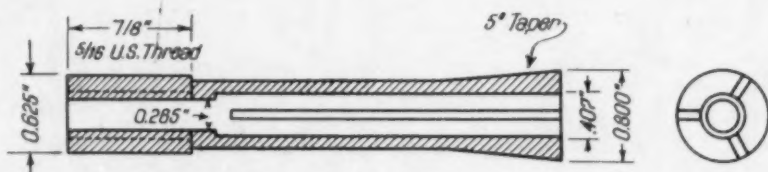
Fig. 2—Diagrammatic Representation of Modified Grip for the Farmer Fatigue Testing Machine.

ing fatigue strengths are generally lower than those reported in the literature for tests under similar conditions (2), (5), (7), (Fig. 1).

The grips exert, according to their design, a pressure distribution which is characterized by stress concentrations at three points of the specimen surface, corresponding to the centers of each grip section. This is proved by a slight wear of the grips at the respective points. The fractures of the specimens therefore show generally a threefold start of the fatigue crack, each part developing into a distinct region, (Fig. 3).

Part I. The Chafing Fatigue Strength of Some Wrought Alloys in Various Conditions*

Materials—A number of wrought metals and alloys have been selected for this investigation with the purpose of covering a broad field. The compositions, treatments, and some properties of these materials are assembled in Table I. The listed alloys comprise one commercially pure metal (aluminum), several alloys of the solid solution type, several precipitation hardenable alloys, a carbon steel, and two austenitic stainless steels.

The metals were received as straight rods 0.400 or 0.375 inch in diameter. Specimens 4 inches long, were cut off and usually subjected to the tests without any further preparation. In the case of duralumin, however, either the metal as supplied or specimens with a surface ground down by 0.001 to 0.002 inch in diameter were used without showing any differences in the results.

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*Abstracted from a thesis presented by P. Stefan in the partial fulfillment of the requirements for the B. S. degree.

Table I

Designation and Composition	Condition	Treatment	Tensile Properties				Hardness	
			PSI Ult. Str.	PSI Yield Pt.	Elon. in 2 In., %	% Red. Area	Rockwell B	Brinell 3000 KG., 10 mm. ball
S.A.E. 1035 Steel	Cold-Worked*	Drawn from ⅞", hot rolled to ⅜", Straightened	105,800	79,800	48.5	16	95	(210)
0.35% C, 0.77% Mn, 0.022% P, 0.028% S, 0.160% Si	Stress-Relieved	Heated to 1150° F. (625° C.), Furnace cooled	....	....	....	....	91	(190)
	Process Annealed	2 hr. at 1300° F. (700° C.), Furnace cooled	....	....	....	....	65	(116)
	Normalized	½ hr. at 1560° F. (850° C.), Air cooled	....	....	....	....	75	(137)
								1000 KG., 10 mm. ball
Stainless Steel No. 1	Cold-Worked*	Drawn from ½", annealed to ⅜", straightened	174,700	104,800	20.0	68.3	40	334
0.05% C, 0.64% Mn, 18.11% Cr, 8.90% Ni, 0.007% P, 0.008% S, 0.540% Si	Annealed*	Cold-worked, annealed, and straightened	107,300	70,200	59.0	74.4	89	166
	Cold-Worked and tempered	Cold-worked, annealed 24 hr. at 1200° F. (650° C.)	....	....	....	....	91	(190)
	Annealed and tempered	Annealed material re-annealed 24 hr. at 1200° F. (650° C.)	....	....	....	....	93	(200)

*Material received from manufacturer in the as tested condition.

(Brinell Hardness in parentheses obtained from conversion chart.)

Table I—(Continued)

Designation and Composition	Condition	Treatment	Tensile Properties				Hardness	
			PSI Ult. Str.	PSI Yield Pt.	Elon. in 2 In., %	% Red. Area	Rockwell B	Brinell 1000 KG., 10 mm. ball
Stainless Steel No. 2 0.06% C, 0.47% Mn, 19.50% Cr, 11.45% Ni, 0.007% P, 0.011% S, 0.235% Si	Cold-Worked	Drawn from 0.400 in. to 0.375 in. Centerless ground	....	....	....	....	95	(210)
	Annealed	Annealed ½ Hr. at 1800° F. (1000° C.), Quenched in cold water	....	....	....	....	77	(141)
	Tempered	Annealed (as above) and tempered 24 Hr. at 1200° F. (650° C.)	....	....	....	....	76	(139)
Iron and Zinc Bearing Aluminum Bronze 87% Cu, 4% Al, 2% Fe, 7% Zn	Cold-Worked* (Commercial Drill Temper)	Drawn from 0.500 in. to 0.380 in., straightened	108,000	59,200	15	....	94	500 KG., 10 mm. ball (171)
	Stress-Relieved*	Drawn, Annealed 30 min. at 550° F. (300° C.), straightened	114,000	81,500	12.5	....	98	(189)
	Annealed*	Drawn, Annealed 70 min. at 860° F. (450° C.), straightened	75,900	46,100	38.5	....	77	(124)
Nickel-Aluminum Bronze 90.5% Cu, 1.5% Al, 8% Ni	Solution Heat-Treated*	1 Hr. at 1650° F. (900° C.), quenched, straightened	48,200	....	37.5	....	38	(73)
	Age Hardened	Solution heat treated, Aged 2 Hrs. at 900° F. (500° C.)	....	....	....	....	83	(137)
	Cold-Worked* and Aged	1 Hr. at 1650° F. (900° C.) on 0.437 in., Quenched, Drawn to 0.375 in., Aged 10 hr. at 860° F. (450° C.)	106,000	....	....	....	95	(175)
Nickel-Aluminum Bronze	Annealed	½ Hr. at 1530° F. (830° C.), ½ Hr. at 1480° F. (800° C.), ½ Hr. at 1300° F. (700° C.), Furnace Cooled	....	....	....	....	87	(148)

*Material received from manufacturer in the as tested condition.

Table I—(Continued)

Designation and Composition	Condition	Treatment	Tensile Properties				Hardness	
			PSI Ult. Str.	PSI Yield Pt.	Elon. in 2 In., %	% Red. Area	Rockwell B	Brinell 500 KG., 10 mm. ball
Duralumin (17S) 95% Al, 4% Cu, 0.5% Mn, 0.5% Mg	Age Hardened* (as received)	Commercial treatment	(60,000)	37,000	22)**	....	67	(106)
	Tempered	1 hr. at 350° F. (175° C.)	......	......	....	....	61	(96)
	Tempered	2 hr. at 450° F. (225° C.)	......	......	....	....	60	(95)
	Annealed	2 hr. at 650° F. (350° C.)	(26,000)	10,000	22)**	....	..	55
	Quenched in Cold Water	Quenched from 950° F. (500° C.) in cold water at 20° C.	......	......	....	....	56	(90)
	Quenched in Boiling Water	Quenched from 950° F. (500° C.) in boiling water	......	......	....	....	55	(89)
Aluminum Alloy (2S) 99.5% Al	Cold-Worked*	Commercial treatment	......	......	....	....	..	25
Aluminum Alloy (3S) 98% Al, 1.2% Mn	Cold-Worked*	Commercial treatment	......	......	....	....	..	42
Aluminum Alloy 52S 97% Al, 2.5% Mg, 0.25% Cr	Cold-Worked*	Commercial treatment	......	......	....	....	..	42
	Annealed	2 Hr. at 650° F. (350° C.)	......	......	....	....	..	69

*Material received from manufacturer in the as tested condition.   **From "Alcoa Aluminum and Its Alloys"—Handbook.
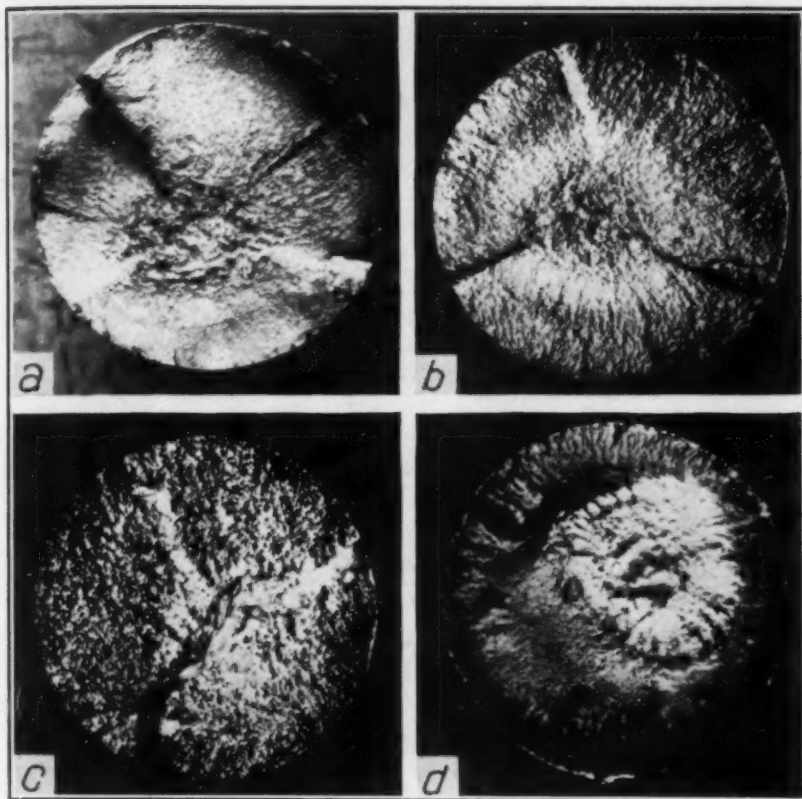
Fig. 3—Macrophotographs of Some Characteristic Chafing Fatigue
Breaks. × 4.
   Fig. 3a—Carbon Steel, Process Annealed. 25,600 Psi.; 570,000 Cycles.
   Fig. 3b—Stainless Steel (No. 2), Annealed. 29,200 Psi.; 772,000
Cycles.
   Fig. 3c—Nickel-aluminum bronze. Solution Heat Treated. 11,700
Psi.; 3,020,000 Cycles.
   Fig. 3d—Aluminum, Cold Worked. 7750 Psi.; 2,003,000 Cycles.

The Behavior of the Different Metals in Chafing Fatigue

Carbon Steel, S.A.E. 1035—(Fig. 4)—The carbon steel investigated was supplied in a cold-worked condition (cold drawn by 15 per cent reduction in area from hot-rolled rod). Specimens were subjected to different heat treatments, such as a stress relief by heating to a temperature of 1150 degrees Fahr. (625 degrees Cent.), a process anneal at 1300 degrees Fahr. (700 degrees Cent.) for 2 hours, and a normalizing at 1560 degrees Fahr. (850 degrees Cent.) for thirty minutes. The normalize slightly reduced the chafing fatigue strength of the cold-worked steel, 21,500 to 20,500 pounds per square inch, while the process anneal resulted in a somewhat larger reduction to a value of 19,500 pounds per square inch. The stress relief treatment, however, gave the rather high chafing fatigue

strength of 25,500 pounds per square inch, about 20 per cent above that of the cold-worked metal.

It is doubtful whether these results fully represent the characteristic behavior of a steel. The initial cold-worked condition is rather poorly defined, and more extensive experiments will be needed to reveal the relations between the chafing fatigue strength and the structural state of a steel, which latter may be varied in such a manifold manner.

Fig. 4—Stress-Cycle Curves in Chafing Fatigue for a 0.35 Per Cent Carbon Steel in Various Conditions.

The absolute values determined by this investigation for all the conditions of steel, except the stress relieved specimens, are considerably lower than those reported by Gillet and Mack (2) (Fig. 1) while a still lower chafing fatigue strength limit (15,000 pounds per square inch) has been reported by Buckwalter and Horger for large specimens (2-inch diameter) of a 0.45 per cent carbon steel in the normalized and drawn condition (6).

Austenitic Stainless Steel, 18-8—(Figs. 5 and 6)—Two different lots of the regular "18-8" austenitic stainless steel (18 per cent chromium, 8 per cent nickel) were investigated, which behaved very differently.

One lot, that which had been cold drawn by 64 per cent reduction in area (and subsequently straightened) had a chafing fatigue strength of 22,500 pounds per square inch. (Fig. 5). Commercial annealing (followed by quenching and straightening) raised the fatigue strength to 27,000 pounds per square inch.

The second lot, which had been cold drawn by approximately 10 per cent reduction in area, straightened and finished by center-

Fig. 5—Stress-Cycle Curves in Chafing Fatigue for the
Stainless Steel No. 1 in Various Conditions.
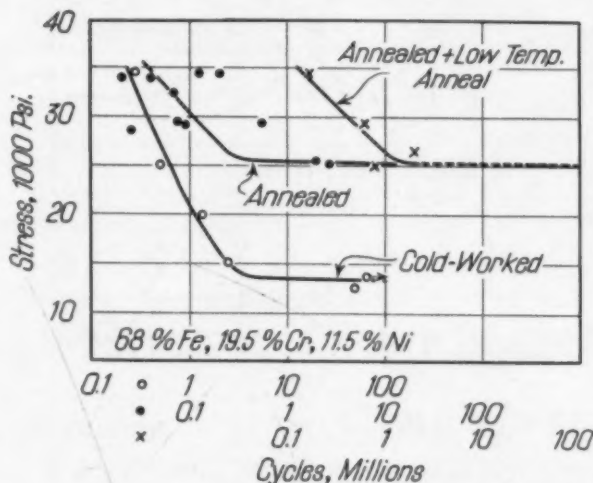
Fig. 6—Stress-Cycle Curves in Chafing Fatigue
for the Stainless Steel No. 2 in Various Conditions.

less grinding yielded the extremely low chafing fatigue strength of
13,000 pounds per square inch (Fig. 6). After annealing at a tem-
perature of 1832 degrees Fahr. (1000 degrees Cent.) for 30 minutes
(in the laboratory) the fatigue strength was raised to well above
25,000 pounds per square inch. The individual results were very
consistent in the case of the cold-worked metal, while considerable
scattering was observed in the annealed condition.

The two austenitic steels differed considerably in their chemical
analysis, the first lot containing about 18 per cent chromium and 9
per cent nickel, the second, 19 per cent chromium and 11 per cent
nickel. This difference, however, in no way explains the peculiar
behavior of the second lot; it might rather have been expected that
the less stable steel, low in nickel, is susceptible to some sort of break-
down in fatigue testing.

The stainless steels have been also subjected to a low temperature anneal at a temperature of 1200 degrees Fahr. (650 degrees Cent.) for 24 hours. Such a treatment causes carbide precipitation and renders annealed steels subject to intercrystalline corrosion. The low temperature anneal, however, little affected the chafing fatigue strength (Figs. 5 and 6) of the annealed condition, while it raised that of the cold-worked metal to a similar extent as a full anneal.
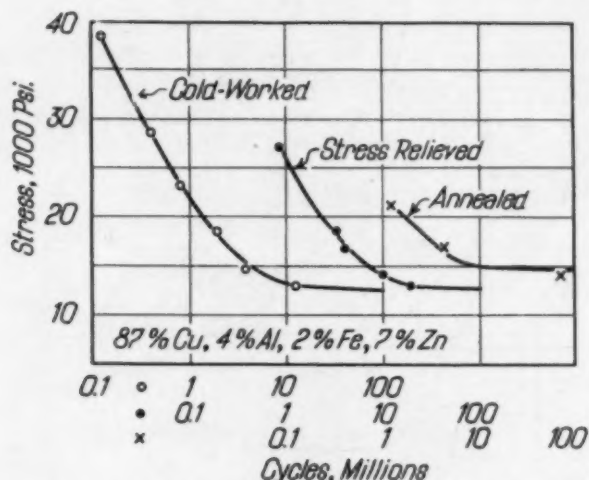
Fig. 7—Stress-Cycle Curves in Chafing Fatigue
for Aluminum Bronze in Various Conditions.

Aluminum Bronze—(Fig. 7)—The aluminum bronze (87 per cent copper, 4 per cent aluminum, 2 per cent iron, 7 per cent zinc) investigated was of the simple type, the aluminum and zinc contents being within the limits of solid solubility and the iron being a favorite constituent for refining the grain and increasing the strength (10).

The chafing fatigue strength of the cold-worked (and straightened) metal has been determined to be 12,000 pounds per square inch, while the annealed (and straightened) condition had a fatigue strength of 14,500 pounds per square inch. A stress relief treatment of the cold-worked metal produced a slight increase in the chafing fatigue strength of the cold-worked metal to 12,500 pounds per square inch.

Heat-treatable Nickel-Aluminum Bronze—(Fig. 8)—Simultaneous additions of nickel and aluminum render copper susceptible to precipitation hardening, and considerable increases in yield point and ultimate strength may be obtained by heat treating alloys of this type (Table I).

Various conditions of such an alloy (90.5 per cent copper, 8 per

cent nickel, 1.5 per cent aluminum) have been investigated. The quenched (and straightened) but not aged metal had a very low chafing fatigue strength, around 10,500 pounds per square inch, while the age hardened conditions, both with and without cold working be-
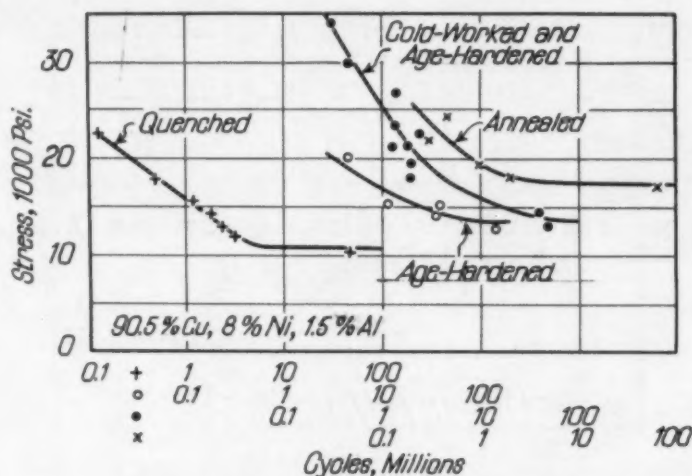
Fig. 8—Stress-Cycle Curves in Chafing Fatigue for a Heat Treatable Nickel-Aluminum Bronze in Various Conditions.
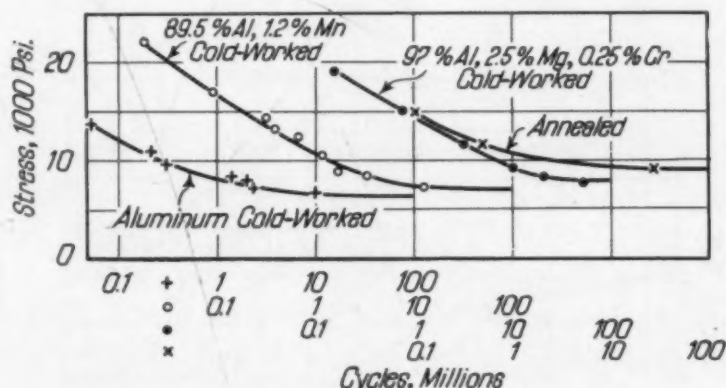
Fig. 9—Stress-Cycle Curves in Chafing Fatigue for Some Aluminum Alloys.

tween quenching and aging (and subsequent straightening) reached a fatigue strength of 13,500 pounds per square inch. An attempt was made to anneal this very sluggish alloy by a slow interrupted furnace cool from 1500 degrees Fahr. (830 degrees Cent.) with a resulting considerable increase in the chafing fatigue strength to a value of 17,500 pounds per square inch, while the hardness of the alloy after such a treatment differed little from that of the age hardened condition.

Simple Type Aluminum Alloys—(*Fig. 9*)—The three different

aluminum-base alloys of the simple (not age-hardenable) type select-ed for this investigation possess a regular fatigue strength within wide limits. They were commercially pure aluminum (2S), an aluminum-manganese alloy (3S, containing 1.2 per cent manganese),
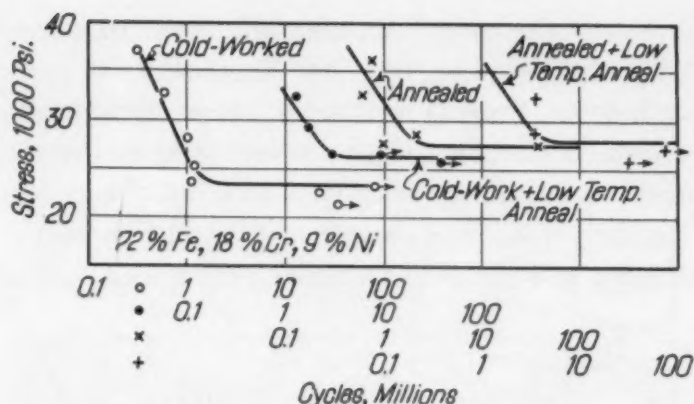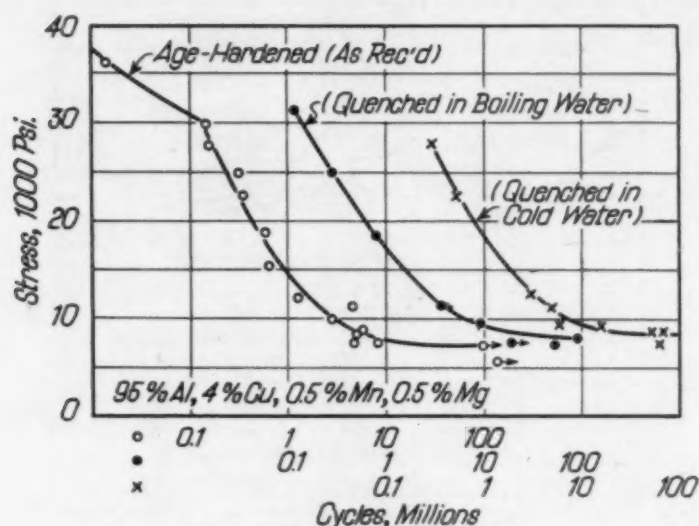

Fig. 10—Stress-Cycle Curves in Chafing Fatigue for Duralumin in Various Conditions.
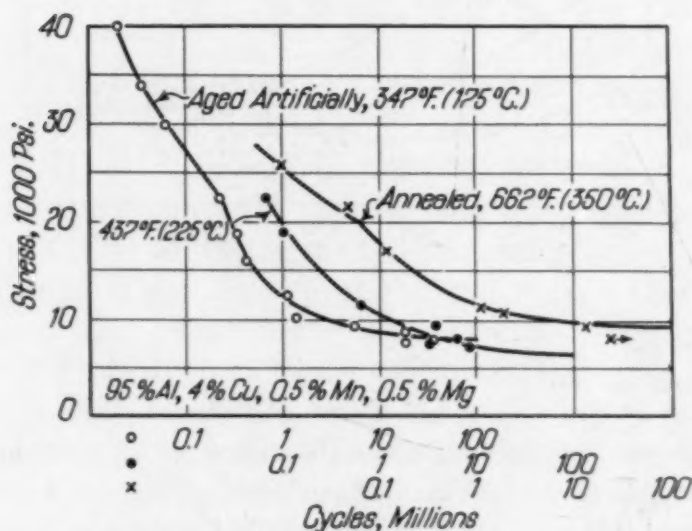
Fig. 11—Stress-Cycle Curves in Chafing Fatigue for Duralumin in Various Conditions.

and an aluminum-magnesium-chromium alloy (52S, containing 2.5 per cent magnesium, 0.25 per cent chromium).

The chafing fatigue strength of the alloys in the as-received cold drawn condition varied little, being 6500, 7000 and 8000 pounds per

square inch respectively. The latter alloy (52S) was also subjected to an anneal that considerably improved the chafing fatigue strength yielding a value of 9000 pounds per square inch.

Duralumin—(Figs. 10 and 11)—The classical heat-treatable aluminum alloy duralumin was investigated in various common and less common heat treated conditions. The age hardened condition (and straightened) as-received gave a chafing fatigue strength of 7500 pounds per square inch. Repetition of the heat treatment in the laboratory raised this value to 8500 pounds per square inch, and quenching from the soaking temperature in boiling water gave an intermediate fatigue strength value of 8000 pounds per square inch.

Aging at 350 degrees Fahr. (175 degrees Cent.) of the as-received metal raised the chafing fatigue strength to 8000 pounds per square inch, while overaging at 440 degrees Fahr. (225 degrees Cent.) lowered the value to 6500 pounds per square inch. A full anneal at 670 degrees Fahr. (350 degrees Cent.) caused an increase to the highest value of about 9000 pounds per square inch observed for any condition of this alloy.

The Effects of Different Treatments on the Chafing Fatigue Strength

Little attention has been paid up to the present time to the effects of mechanical and thermal treatments on the fatigue strength in the presence of stress raisers. Contrary to the regular fatigue strength, the chafing fatigue strength usually does not follow the trend of the tensile strength. It is generally believed that any of the stress raisers reduces the fatigue strength to a low level which is almost constant for a certain group of alloys, such as the carbon steels (Fig. 1), and which is then mainly determined by the specific type of stress raiser. Only an irregular scattering within rather narrow limits has been observed for the commercial conditions of any specific metal.

The present investigation is not extensive enough to establish definite generalizations regarding the effects of different treatments on the chafing fatigue strength. It has been found, however, as a general trend that the more stable conditions, such as the fully annealed and sometimes the stress relieved states, possess a higher chafing fatigue strength than the stronger cold-worked or heat treated metals.

To reveal such effects the fatigue tests have to be carried out to

high numbers of cycles, to at least 50 millions of revolutions. The course of the stress cycle diagrams at higher stresses corresponds generally to the magnitude of the tensile strength, but at low stresses approaching the chafing fatigue limit the curves for the different conditions of a specific metal as a rule intersect each other.

Thus most of the investigated alloys exhibit a higher chafing fatigue strength in an annealed than in a cold-worked condition. This difference amounts to 10 per cent for the tested aluminum alloy (Fig. 9) and 20 per cent for the aluminum bronze (Fig. 7). The two lots of stainless steel show an increase by 20 and 95 per cent, respectively (Figs. 5 and 6), of the chafing fatigue strength of the cold-worked specimens on annealing. Only with the carbon steel, a slight decrease of this fatigue limit was found on normalizing the cold-worked specimens; however, an incomplete anneal, which has been designated as a stress relief, also brought about a 20 per cent increase in the chafing fatigue strength. It can be therefore concluded that the increase in hardness and strength caused by cold-work is not accompanied by a corresponding increase of the fatigue strength if stress raisers are present.

However, this effect of cold work is not fully explained as yet, and to reach this purpose a reverse procedure should be applied from that followed in this investigation, namely, a definite cold work subsequent to a definite anneal rather than an anneal applied to a not completely defined cold-worked condition.

The heat treatment of alloys which are susceptible to precipitation hardening, associated with substantial increases in the tensile strength and the regular fatigue strength, reduces on the contrary the chafing fatigue strength compared with that of the annealed condition. Thirty per cent difference has been observed for these two conditions of the nickel-aluminum bronze (Fig. 8) and 20 and 5 per cent, respectively, for duralumin (Figs. 10 and 11), of which two slightly different age hardened lots (as received and age hardened in the laboratory) have been investigated.

Particularly low chafing fatigue limits have been observed for these precipitation-hardenable alloys if subjected to special heat treatments, such as the solution heat treatment of the nickel-aluminum bronze, (Fig. 8) and an overaging of the duralumin (Fig. 11). It may be further remembered that the one lot of austenitic stainless steels, which also has a certain tendency to precipitate, showed an extremely low chafing fatigue limit in the cold-worked condition. It

appears, therefore, that alloys which are susceptible to precipitation may develop a particular sensitivity to stress raisers under certain conditions which, however, cannot as yet be explained.

It was expected that this sensitivity may be correlated to the tendency of some alloys of this type for developing susceptibility to intercrystalline corrosion if heat treated in a specific manner. Such alloys are the austenitic stainless steels and the duralumin; and they have been therefore heat treated within temperature ranges in which such a tendency to intercrystalline cracking develops (Table I). The chafing fatigue strength of both alloys, however, after such heat treatments was found to be a little different from the condition to which this heat treatment was applied if no substantial changes of the mechanical properties occurred.

Cold work after quenching but prior to aging had no visible effect on the chafing fatigue strength of the nickel-aluminum bronze (Fig. 8).

THE EFFECT OF RESIDUAL STRESSES ON THE CHAFING FATIGUE STRENGTH

A controversial problem is the previously discussed increase of the fatigue strength in the presence of stress raisers caused by a mechanical surface treatment, such as surface rolling. According to our present knowledge, this treatment introduces two definite changes of the metal: (a) Strain hardening of the surface layers (5), and (b), a development of high residual compression stresses in the surface layers (8). Either one of these two components of surface-rolling has been made responsible for the beneficial effect of the mechanical surface treatment.

The present investigation permits evaluation of the effects of each of these two components of a mechanical surface treatment without interference of the other.

The cold-work itself has been usually found to reduce the chafing fatigue strength, as discussed in the previous section.

Residual stresses may be developed in commercial manufacturing either of the compression type or of the tension type. Quenching from high temperatures develops compression stresses in the surface layers. Such quenching stresses have been found to somewhat improve the fatigue strength of spring steels which is usually low because of the presence of stress raising irregularities on the surface (13).

However, this effect is much smaller than that obtained on surface rolling.

The effect of residual compression stresses on the chafing fatigue strength can be evaluated from experiments with stainless steel and duralumin, which have been investigated in otherwise very similar conditions, but differing by the presence or absence of quenching stresses, (Figs. 5 and 6). The quenched (after annealing) stainless steel has been found to possess a slightly lower chafing fatigue strength than the same steel subjected to an additional anneal at 1250 degrees Fahr. (680 degrees Cent.) which presumably eliminates the major portion of the quenching stresses. Duralumin, quenched in cold water (Fig. 10), on the other hand, possesses a slightly higher chafing fatigue strength than in an almost stress-free condition, obtained by quenching in boiling water (14). Thus, contradictory results have been obtained on this point; but the effect is in both cases small and again in no way comparable to the substantial improvement obtained by surface rolling.

Therefore, the actual source of the beneficial effect of surface treatments on the fatigue strength in presence of stress raisers cannot be explained as yet. Both strain hardening and residual stresses of the compression type have only a minor effect on the chafing fatigue strength.

Residual stresses of the tension type are generally present in cold drawn rods and wire. However, the metals used for this investigation have been received from the suppliers as straightened rods and may therefore contain compression stresses in the surface layers. Residual stresses in cold-worked metal can be also considerably reduced by a stress relief anneal without otherwise affecting the properties (hardness, tensile strength, etc.) of the metal. Two alloys have been investigated in both a cold drawn and a stress relieved state, carbon steel (Fig. 4) and aluminum bronze (Fig. 7); and in both instances an increase of the chafing fatigue strength has been effected by stress relieving the cold-worked metal. This effect, also, cannot be explained. Residual tensile stresses should exert little influence in the presence of other more powerful stress raisers.

REGULAR FATIGUE STRENGTH AND CHAFING FATIGUE STRENGTH

The relation of the chafing fatigue strength to the regular fatigue strength has been previously discussed for steel, (Fig. 1). The

fatigue strength of low strength soft steels is little affected by chafing, while it is reduced to a fraction in the case of high strength steels. Thus, little benefit can be derived from alloying and heat treatment regarding the chafing fatigue strength of steels.

Similar results have been obtained in this investigation for alu-
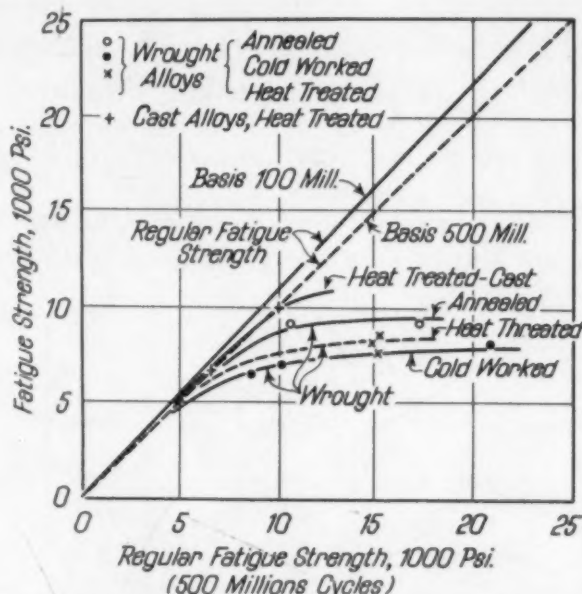
Fig. 12—Relation Between the Regular Fatigue Strength and the Chafing Fatigue Strength of Various Aluminum Alloys.

minum alloys. In Fig. 12 the chafing fatigue strength of aluminum and aluminum alloys is plotted over the regular fatigue strength, as given in the publication "Alcoa Aluminum and Its Alloys" (11) on the basis of 500 million cycles. Lines representing the regular fatigue strength for a basis of 500 and 100 millions, respectively, have been drawn to facilitate the comparison between the regular and the chafing fatigue strength. It has to be remembered that the latter was determined on a basis of 100 million cycles.

According to Fig. 12, the effect of chafing on the fatigue strength depends to a considerable extent upon the condition of the metal. The hard conditions, both strain hardened and precipitation hardened, are suffering in their fatigue strength more by chafing than the annealed state. Low fatigue strength values are generally little affected by chafing. On the contrary, chafing affects alloys with high regular fatigue strengths to such an extent that, in the case of the cold-worked alloys, the chafing fatigue strength may be as low as 35 per cent of the regular fatigue strength. Thus, for the cold-worked alloys the

chafing fatigue strength appears to be limited to a maximum value of 8000 pounds per square inch, while the annealed conditions approach a maximum chafing fatigue of about 9500 pounds per square inch. The heat treated duralumin and another heat treated alloy, discussed in Part II, determine the maximum chafing fatigue strength of heat treated alloys to about 8500 pounds per square inch. Cast alloys which will also be discussed later have been included in Fig. 12 and appear particularly favorable regarding the effect of chafing.

Conclusions

The investigation of a number of different wrought metals in different conditions shows that the endurance testing of cylindrical
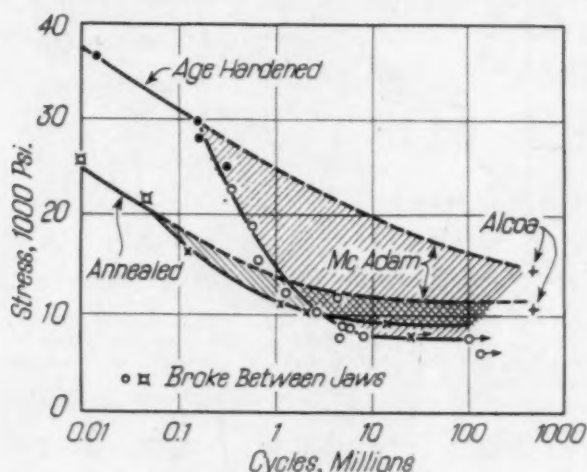
Fig. 13—The Effect of Chafing on the Stress-Cycle Curves of Duralumin in the Annealed and Heat Treated Conditions.

specimens in reverse bending yields uniform and reproducible results. This method reveals the behavior and fatigue strength of the metal subjected to the simultaneous attack of vibrating stresses and lateral pressure, or "chafing," such as present in press fits, axle seats, and propeller hubs.

It has been made probable that the nature of such chafing consists in a particular type of stress raising, related to that of notches and of corrosion on fatigue. Generally the effect of stress raisers is more pronounced at low stresses which produce the break after a long time or many millions of cycles than with high stresses; and stress raisers have often no effect at all if a break occurs before less than 100,000 cycles (Fig. 13).

The effect of stress raisers increases with the increase of the

Table II

Designation and Composition	Condition	Treatment	Tensile Properties			Brinell Hardness	
			Ult. Str. (Psi)	Yield Str. (Psi)	Elon. in 2"—%	3000 kg. Load 10 mm. Ball	500 kg. Load 10 mm. Ball
Steel 0.24% C, 0.43% Si, 0.75% Mn, 0.03% P, 0.03% S	Cast and Normalized	3 Hours at 1625° F. (885° C.), Air Cooled	76,600	......	30	128	...
	Forged and Normalized	1 Hour at 1625° F. (885° C.), Air Cooled	80,000	56,000	28**	131	...
Aluminum Alloy (195 T 4) 96% Al, 4% Cu	Sand Cast and Heat Treated	12-14 Hours at 960° F. (515° C.) and Quenched in Boiling Water	31,000*	16,000*	8.5*	...	...
Aluminum Alloy (B 195 T 4) 92.5% Al, 4.5% Cu, 3% Si	Chill Cast and Heat Treated	6 Hours at 950° F. (510° C.) and Quenched in Cold Water	36,000*	22,000*	7.5*	...	...
Aluminum Alloy (25 S) 93.9% Al, 4.5% Cu, 0.8% Si, 0.08% Mn	Forged and Heat Treated	10 Hours at 960° F. (515° C.), Quenched in Cold Water, and aged 10 Hours at 340° F. 170° C.)	55,000*	30,000*	16.0*	...	112

*Alcoa Aluminum and its Alloys (1938).
**Broke in Jaw.

regular fatigue strength, in such a manner that a specific type of alloys, such as ferritic steels, or aluminum alloys, possesses an almost constant fatigue strength under specific conditions where a severe stress raiser is present.

This limiting value is generally higher for the annealed than for other conditions, such as strain hardened, solid solution heat treated

and precipitation hardened metal. It appears that the chafing fatigue strength of a specific metal reaches the highest possible value in the most stable condition.

Residual stresses introduced by both cold work and quenching exert only little effect on the chafing fatigue strength.

Part II

THE CHAFING FATIGUE STRENGTH OF CAST AND WROUGHT ALLOYS OF SIMILAR COMPOSITIONS*

Materials—For the comparison of the chafing fatigue strength of cast and wrought alloys, two alloy types have been selected, a 0.25 per cent carbon steel and some aluminum alloys with a copper content of 4 to 5 per cent. The compositions, treatments, and some properties of these materials are assembled in Table II.

The metals were received in different shapes, heat treated in a manner corresponding to their commercial condition, and then machined and finished. For the chafing fatigue experiments, specimens of 0.375 ($\pm$ 0.005) inch diameter were used for the steels and of 0.400 ($\pm$ 0.005) inch diameter for the aluminum alloys in the as-turned condition without further surface treatment.

Regular fatigue strength curves were also determined using the common necked and polished specimens with a minimum diameter of 0.40 inch and a radius of 9.84 inch.

The steel was received in the form of an ingot, about 6 x 6 x 30 inches, and a number of wedges, about 6.5 inches long and 5.5 inches high, ¾ inch wide at the base and 2.5 inch wide at the top. A part of the ingot was broken down by forging into rods approximately ½ inch diameter, while two fatigue specimens were cut from the base part of each wedge. The sand cast aluminum alloy was supplied in rods, ½ inch in diameter and 30 inches long, the chill cast aluminum alloy in the usual tensile test bars, ½ inch in diameter, and the wrought aluminum alloy in forged bars, ½ inch in diameter.

THE FATIGUE CHARACTERISTICS OF THE DIFFERENT ALLOYS

Steel—(Fig. 14). The plain carbon cast steel which was investigated showed a reduction in the fatigue strength by chafing

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*Abstracted from a thesis presented by E. J. Jory in the partial fulfillment of the requirements of the B. S. degree.

of 30 per cent. The values were 34,500 pounds per square inch for the regular fatigue strength and 25,000 pounds per square inch for the chafing fatigue strength. In running the regular fatigue specimens considerable irregularity was noted in the lower regions of the stress-cycle curve.
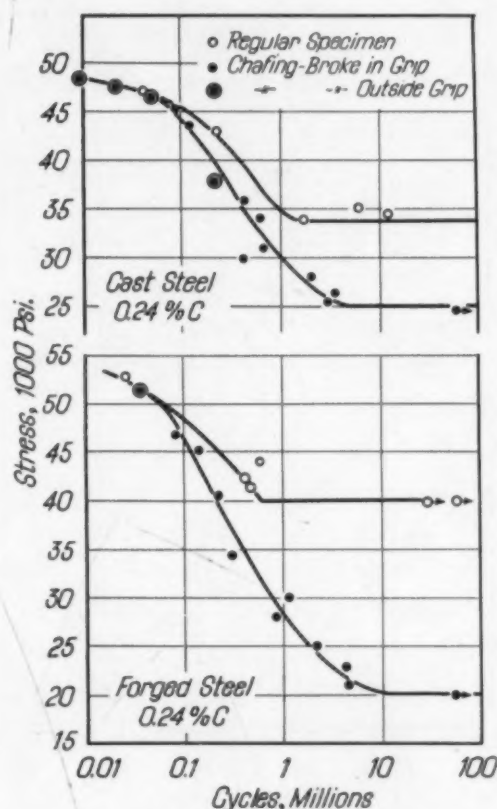
Fig. 14—Stress-Cycle Curves in Regular and Chafing Fatigue for a Forged and a Cast 0.24 Per Cent Carbon Steel in the Normalized Condition.

The forged steel investigated was of the same composition as the cast steel and received the same type of heat treatment. The effect of chafing was more noticeable in the forged steel than in the cast steel in that a 50 per cent reduction in the fatigue strength was observed. The values were 40,000 pounds per square inch for the regular fatigue strength and 20,000 pounds per square inch for the chafing fatigue strength.

Thus, while the regular fatigue strength of the forged steel is greater than that of the cast steel, the chafing fatigue strength is less by 20 per cent than that of the cast steel.

From the stress-cycle curves for the two steels it is seen that

for both types of fatigue the uppermost portions of the curves (high stresses and low numbers of cycles) coincide. In this portion of the curves the chafing fatigue specimens broke generally outside the grips showing that there was no strength reducing effect of chafing under such conditions.
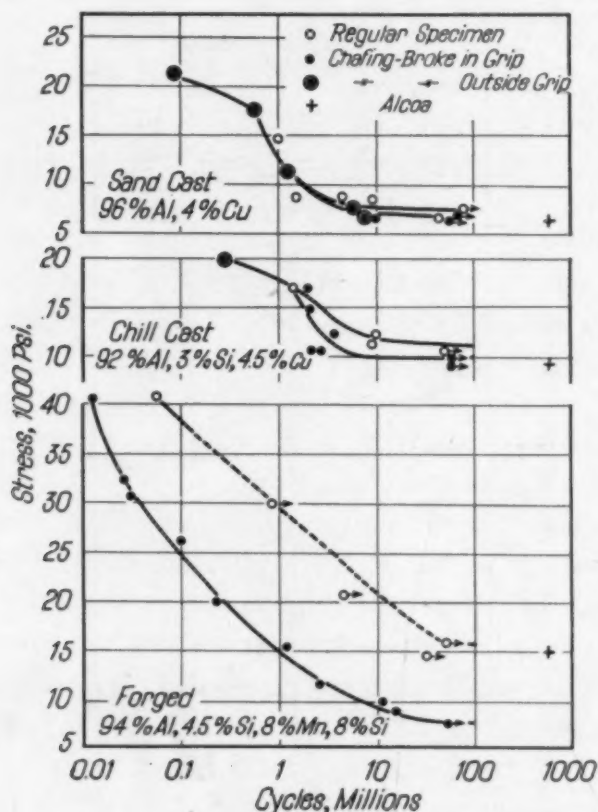
Fig. 15—Stress-Cycle Curves in Regular and Chafing Fatigue for a Forged and Two Cast Aluminum-Copper Alloys in the Heat Treated Conditions.

Aluminum Alloys—(Fig. 15). The sand cast aluminum alloy which was investigated showed that there was little effect of chafing, less than 10 per cent reduction due to chafing being observed. The fatigue limit in chafing fatigue was about 6500 pounds per square inch and in regular fatigue was 7000 pounds per square inch. Correspondingly, only one chafing fatigue specimen which was subjected to a stress very close to the fatigue limit broke in the grip, revealing the effect of chafing. All other chafing specimens broke outside the grip.

The chill cast aluminum alloy also showed a small reduction in fatigue strength due to chafing. The regular fatigue strength was

11,000 pounds per square inch, while the chafing fatigue strength was 10,000 pounds per square inch. However, with this alloy, the small difference is sufficient to produce a definite chafing break at all stresses under 20,000 pounds per square inch. The different behavior of the sand cast and the chill cast alloy, in this respect, is difficult to explain.

The wrought aluminum alloy investigated showed a reduction in fatigue strength of 50 per cent caused by chafing. The regular fatigue strength was approximately 16,000 pounds per square inch and the chafing fatigue strength was 8000 pounds per square inch. In determining the regular stress-cycle curve for the wrought aluminum alloy some difficulty was encountered in that the necked and polished specimens often failed prematurely in the jaws as a result of the chafing effect. This is caused by the large difference btween the stress-cycle curves for the two types of fatigue which exists in an unusual manner over the whole range of the present investigation. A different shape of specimen should be used for alloys which are greatly affected by chafing.

In the case of the aluminum alloys, similar to the steels, the wrought alloy has a higher regular fatigue strength than both cast alloys. The chill cast alloy, however, possesses a definitely higher chafing fatigue strength, by about 25 per cent than the wrought alloy. The sand cast alloy has a regular fatigue strength below the chafing fatigue strength of the wrought alloy but shows little effect of chafing.

Conclusions

It may be concluded from the investigation of the forged and cast steels and aluminum alloys that cast metals are generally less susceptible in fatigue to the effect of chafing. As chafing appears to be closely related to other types of stress raisers it is probable that the fatigue strength of cast metals is also less affected by notches and corrosion than that of wrought metals.

This superiority of the cast metals is illustrated in Fig. 12 for the aluminum alloys. The trend curve for the chafing fatigue strength of the heat treated cast alloys is located about 30 per cent higher than that of the heat treated wrought alloys. However, the actual differences are smaller, because the cast alloys have generally lower regular fatigue strength than the wrought alloys. The sand cast aluminum alloy, in particular shows a lower chafing fatigue

strength and even a lower regular fatigue strength than the chafing fatigue strength of the forged alloy.

It would appear, therefore, that suitably selected cast alloys may offer some advantage for applications where moving machinery parts are exposed to severe surface attacks by notches, corrosion and chafing.

The nature of the discovered effect cannot be explained as yet. It is, however, in line with the conclusions in Part I, that the chafing fatigue strength of wrought metals is usually the highest for the most stable condition. There is also a parallelity apparent between the chafing fatigue strength and the creep strength (at high temperatures), as cast and coarse-grained metals are frequently referred to as having a higher creep resistance than a fine-grained wrought metal.

ACKNOWLEDGMENT

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DISCUSSION

Written Discussion: By O. J. Horger, in charge of railway engineering and research, The Timken Roller Bearing Co., Canton, Ohio.

Chafing or fretting action is the cause of fatigue failure in so many different kinds of machine parts and members that its importance is not generally appreciated. In this respect the authors have added to our knowledge of this subject. Several questions arise, however, with reference to their paper as follows:

Chafing tests made by the discusser have always produced two endurance limit values (1), that is, (a) the value above which incipient fatigue cracks will not develop and (b) usually a larger value than (a) above which actual breaking off of the specimen occurs, both of which values are established after about 100 million revolutions. Various materials, heat treatment, and design of specimens are some factors which determine the values for (a) and (b). Under some conditions there may be only a small or no difference between the values for (a) and (b) or under other conditions this difference may be very large. This existence of double endurance values, under chafing conditions only, is not recognized by the authors as apparently only values for (b) are given. Determination of value (a) requires magnetic examination of the fatigue specimens and/or microscopic methods and close observation in view of the oxide products of rubbing corrosion interfering with determining the presence of small fatigue cracks. Have the authors made similar observations?

Examples of above endurance limit values have been previously published* as follows:

Kind of Axle Wheel Seat Treatment — Same Axle Design and S.A.E. 1045 Steel in All Cases.	Endurance Limit in Wheel Seat of 2-Inch Axle (Where Chafing Takes Place) to prevent:	
	Initiation of Fatigue Cracks—(a) Value	Breaking off of Axle—(b) Value
Normalized and Tempered	8,000	13,000
Surface Cold Worked	8,000	35,000
Metallized	21,000	21,000
Flame Hardened	25,000	45,000

*See footnotes 1, 2, 3 on page 400.

and indicate very low values for (a) which are much lower than obtained in notch fatigue tests. On this basis it is difficult to see any correlation between chafing and notch fatigue tests.

Also the discusser finds much scatter in chafing tests and particularly near the endurance limit values. Again as the wear of the grips increases up to a certain amount, it is found that this wear tends to increase the life of subsequent test specimens. For this reason our practice is to use new fitted members or grips with each new specimen tested. In this connection, the authors establish endurance limit values with what we would consider too few points.

Findings of the discusser are contrary to those of the authors, in that the (b) endurance limit may be increased appreciably by surface cold working and certain heat treatments as shown in above table. Alloying also gives some increase.

Unfortunately the (a) value is not always increased with the (b) value. Therefore, if the part is inspected regularly, per airplane and railroad practice, and is scrapped if cracks are found, it becomes necessary for the designer to use the (a) endurance value. On the other hand, if the part is never inspected during its life then the (b) value may be used by the designer.

Oral Discussion

M. D. STONE:[4] Does Dr. Sachs have some idea of what the mechanism of this chafing fatigue is and how it can be mitigated by lubricants or what not?

It is a little beside the point, possibly, but something that has concerned us from the design point of view.

Authors' Reply

As far as I understand, Mr. Horger refers to the presence of a "damage line" as introduced by French for the regular fatigue tests. The presence of such a damage line is generally recognized, but it has been little used in any type of fatigue testing up to the present time. The authors have made no attempt to determine the limit at which the metal might be damaged by vibrational stresses and have followed the usual procedure of determining the fatigue limit.

Regarding the scattering of the values, Mr. Horger will find that, with the exception of the stainless steels, surprisingly consistent readings were obtained in our tests. We have used hardened grips and the wear of the grips has been insignificant so far. We realize however that the conditions with much larger specimens, as used by Mr. Horger, may be more complicated in this respect.

[1]T. V. Buckwalter, O. J. Horger and W. C. Sanders, "Locomotive Axle Testing", *Transactions,* American Society of Mechanical Engineers, Vol. 60, 1938, p. 335.

[2]O. J. Horger and T. V. Buckwalter, "Fatigue Strength of 2-Inch Diameter Axles with Surfaces Metal and Flame Hardened", American Society for Testing Materials paper, June 1940.

[3]O. J. Horger and T. V. Buckwalter, "Photoelasticity Applied to Design Problems", *Iron Age,* May 23, 1940.

[4]Manager, Development Department, United Engineering & Foundry Co., Pittsburgh.

The surface cold working and the heat treatments, which, according to Mr. Horger, improve the chafing fatigue limit are not regular treatments, but fundamentally change the stress conditions in the vibrating specimens. There has been considerable argument about the source of this improvement, and several theories have been advanced. None of these theories have been generally acknowledged, but it appears that high compression stresses localized at the surface are one important factor.

Regarding the relation between notch fatigue strength, chafing fatigue strength and corrosion fatigue strength, it appears that chafing ranges somewhere between notching and corrosion. The similarity in the general trend of these three effects is very apparent, while this cannot be explained as yet in a simple manner. However, the general conception of stress concentrations is able to qualitatively explain either effect.

Apparently the main "stress raising" factor is the high pressure perpendicular to the surface. A sudden section change from a heavy to a light section may reduce the fatigue strength of some metals as much as chafing. While the chafing or fretting action appears to be insignificant in some cases, it may include additional corrosion or oxidation in others. In other words, chafing may act either similar to notches and section changes or to corrosion, depending upon the particular metal.

THE ALLOYS OF MANGANESE AND COPPER:
VIBRATION DAMPING CAPACITY

By R. S. Dean, C. Travis Anderson, and E. V. Potter

Abstract

Preliminary measurements showed that certain manganese-copper alloys which possess a remarkable deadness or lack of metallic ring also have unusually high vibration damping capacity at low stresses.

An instrument for the determination of this property is described and the results obtained with it using manganese-copper alloys are discussed.

The results indicate that the high values obtained by certain heat treatments are due to the formation of ordered anti-phase nuclei in the mass of the alloy.

The vibration damping capacity appears to reach a maximum with some critical distribution of these anti-phase nuclei and decrease as a state of equilibrium order is approached. The effect of cold work is to decrease the vibration damping capacity of the alloys as would be expected from its disordering effect. The unusually high vibration damping capacity of the quenched alloys drops with time, and it is suggested that it is due to their failure to come to an equilibrium volume when quenched. There would thus seem to be a similarity in structure with regard to its effect on vibration damping capacity between an alloy in a supercooled state and an alloy in a partly ordered state. In both instances, some of the atoms of the metal are not in the equilibrium positions required by the temperature. High vibration damping capacity appears to be an attribute of such metastable systems.

THE remarkable deadness or lack of metallic ring in certain manganese-copper alloys was noted early in our investigation of this system. Preliminary measurements showed that this lack of metallic ring was associated with an unusually high vibration damping

capacity at low stresses. It was also observed that this unusually high vibration damping capacity was present only in alloys containing a considerable proportion of manganese and that it varied greatly with mechanical and heat treatment. These preliminary results are discussed in an earlier paper.[1]

This paper deals with a more systematic investigation of the vibration damping capacity of these alloys and its relation to composition and treatment.

Preparation of the Alloys

The preparation and method of treatment of the alloys used in this investigation have been described in an earlier publication of this series. It may be pointed out here that "quenched" means quenched from a temperature just below the solidus. Heat treatment always started with the quenched alloys and was done in vacuum for a 2-hour period unless otherwise stated.

Method of Measurement

This paper covers only vibration damping capacity at very low stresses. The method used for measurement was similar to that of Wegel and Walther.[2] This was one of the so-called "forced vibration" methods. The specimen is in the form of a long rod, about ¼ inch in diameter and 10 inches in length, and is supported on fine steel wires, 0.005 inch in diameter, situated so that they will absorb a minimum amount of energy from the specimen. The rod was set into longitudinal vibration by means of an electromagnet (called the driving magnet) coupled to one end, and the amplitude of the vibrations was measured by means of another electromagnet (called the receiving magnet), coupled to the other end.

The procedure is to determine the amplitude of the vibrations for a number of frequencies above and below one or more of the resonant frequencies of the rod. From these data, an amplitude-frequency or resonance curve can be drawn, and all the quantities required to calculate the specific damping capacity can be obtained from it. Throughout this paper, whenever the term "vibration damping capacity" is used, the "specific damping capacity" is meant. This

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[1] R. S. Dean, C. Travis Anderson, Cresap Moss and P. M. Ambrose, "Manganese and Its Alloys," U. S. B. M., RI-3477, Nov. 1939.

[2] R. L. Wegel and H. Walther, "Internal Dissipation in Solids for Small Cyclic Strains," Physics, Vol. 6. April 1933, p. 141.

quantity is the ratio of the energy loss per cycle to the potential energy at maximum stress and is discussed by von Heydekampf.[3]
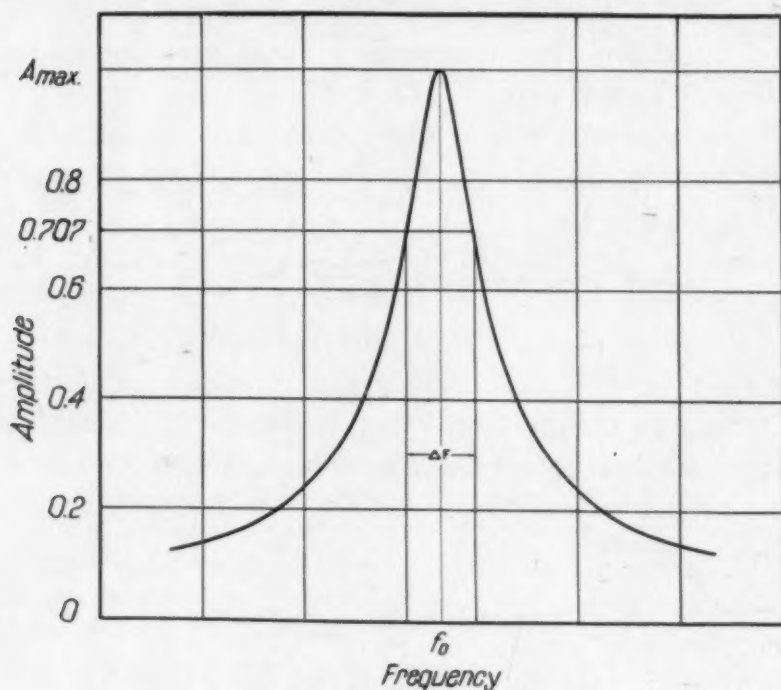
Fig. 1—Typical Resonance Curve.

A typical "resonance curve" is shown in Fig. 1. Without going into the theory any further, the "specific damping capacity," usually referred to as "p", can be expressed as "p" $= 2\pi \dfrac{\Delta f}{fo}$ and usually is expressed in per cent. Referring to Fig. 1, A_{max} is the maximum amplitude of vibration of the specimen, fo is the resonant frequency or the frequency giving the maximum amplitude, and Δf is the width of the curve in cycles per second at an amplitude equal to $\sqrt{2}/2$ or 0.707 times the maximum amplitude. Thus, from the resonance curve all the data required to calculate "p" can be obtained.

In Fig. 2, the driving and receiving magnets and the wire support for the specimen are shown. As all the measurements reported in this paper were made with the rods vibrating at their fundamental or lowest resonant frequency, there was only one node or point of minimum vibration in the rod, and it was at the center of the rod. Accordingly, the wire support was placed at the center of the rod. If, for convenience or any other reason, it would be desirable or

necessary to support the rod at more than one point, it would be necessary to drive the rod at one of the higher-order harmonic resonant frequencies. The number of nodes is equal to the order of the harmonic frequency, two for the second, three for the third, etc. In the case of the second harmonic, the two nodes are a quarter
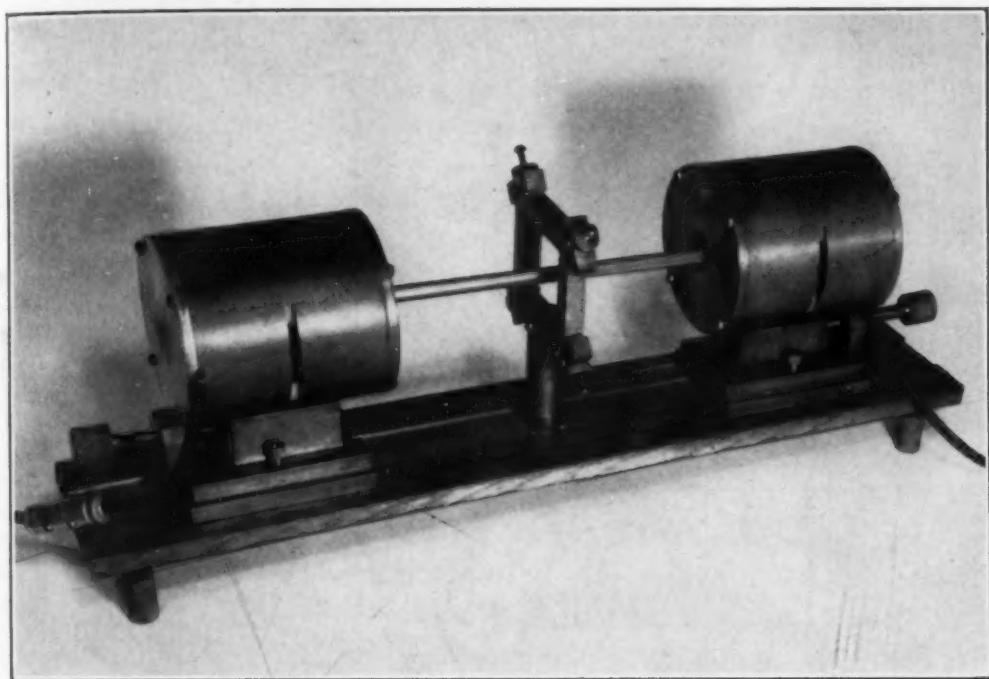
Fig. 2—Driving and Receiving Magnets and Specimen Support for Measuring Vibration Damping Capacity.

of the rod length from each end, and for the third harmonic, two are at points one-sixth of the rod length from the ends, and the third is at the center of the rod. A more complete discussion of the vibration of rods is given in "Theory of Vibrating Systems and Sound," by I. B. Crandall, D. van Nostrand Co., Inc., New York, 1927.

The rod support shown in Fig. 2 consists of a square brass frame, in the corners of which four screws are arranged so that they can be moved in or out along the diagonals of the square frame without actually turning. The adjustment on these screws is such as to make very small movements of them possible, as this greatly facilitates balancing the specimens. Two short lengths of wire are fastened to the inner ends of the screws, forming two loops, one from the top two screws and one from the lower two screws. The rod is balanced by supporting its weight on the upper loop, balancing

it as nearly as possible. The lower loop, passing over the rod and being as close to the upper loop as possible, is then tightened until the small couple exerted by the two loops balances the rod in a horizontal position. The tension in the loops should be as small as possible to minimize the energy loss in the support. The pull of the driving and receiving magnets on the ends of the rods helps considerably in stabilizing the balance of the rods.
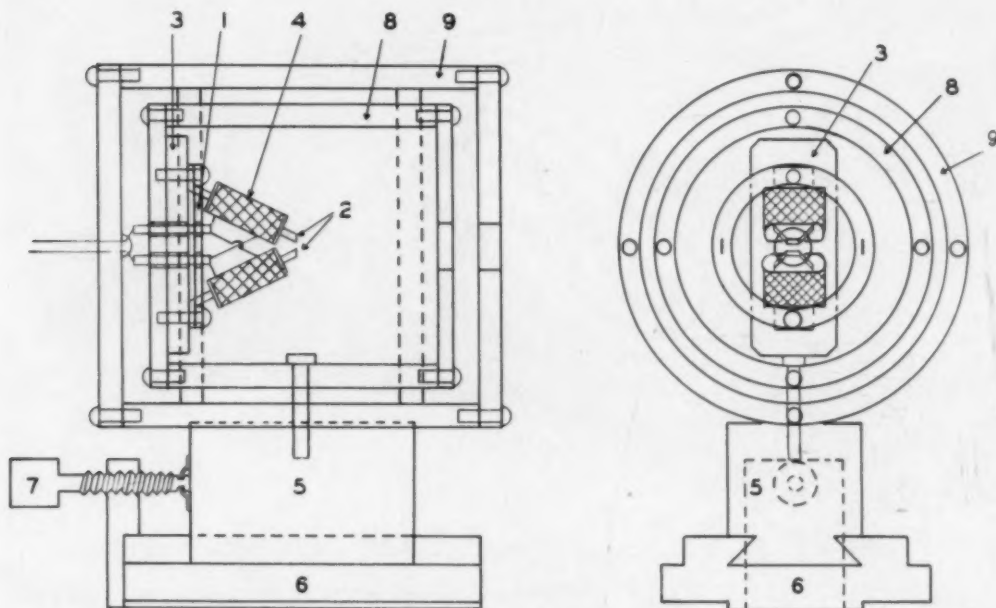
Fig. 3—Cross Section of Magnets.

A cross section of the driving magnet is shown in Fig. 3. The magnets consist essentially of two semicircular permanent magnets 1 and two soft-iron pieces 2, one end of each being clamped between the permanent magnet and a brass bar 3. The coils 4, through which the driving current passes, are wound on small fiber forms on pole pieces 2. The end of the specimen is placed as close to the faces of the pole pieces as possible without having the magnets pull the specimen into contact with the pole faces. This distance is in the order of 0.010 inch. To make this adjustment, the magnets in their housings are mounted on a base 5, which slides along a larger base 6, a screw 7, serving to move one base with reference to the other. A coarse adjustment is made by sliding base 6 on a large brass base 4 inches wide and 22 inches long, holes in this larger base 1 inch apart being used to firmly pin blocks 6 in the nearest position usable. The final fine adjustment is then made by screw 7. Slits in the side of the

magnet housings, visible in Fig. 2, make it possible to see the pole faces and enable one to adjust the spacing between the specimen end and the pole faces more readily.

Permanent magnets 1 supply a constant polarizing field, which helps hold the specimens in balance, and as long as the field due to the magnetizing current in coils A is less than this polarizing field the pull on the ends of the specimens has the same fundamental frequency as the magnetizing current. If this polarizing field were not present or if the field due to the magnetizing current greatly exceeded the polarizing field, the pull on the specimens would have a fundamental frequency double that of the magnetizing current.

The magnet assemblies were fastened in iron cylinders to concentrate the field around them and prevent the field of the driving magnet from acting on the receiving magnet directly. The electrical connections were made through one end of the cylinder, and the specimens entered through a hole in the center of the face opposite the magnet pole faces. It was found that these iron shields were not effective enough and that the direct coupling between the magnets interfered with measurements on high damping materials. This was remedied by rotating the receiver magnet assembly through 90 degrees so that its axis was horizontal and placing the iron cylinders inside larger brass cylinders 9. This combination of brass and iron proved quite effective and, combined with the 90 degrees orientation of the magnets, resulted in no measurable coupling existing between the coils.

The outer shields were of $\frac{1}{4}$ inch brass, 4 inches in diameter and $4\frac{1}{2}$ inches in length. The inner shields were of $\frac{1}{4}$ inch iron, $3\frac{3}{8}$ inches in length and $3\frac{1}{8}$ inches in diameter. The pole pieces were of iron about $\frac{7}{16}$ inch wide and $\frac{1}{16}$ inch thick, the pieces being tapered from $\frac{7}{16}$ inch in the coils to $\frac{1}{8}$ inch at the pole faces. The permanent magnets were obtained from an old telephone head set. The only difference between the two magnets is in the number of turns on the coils. On the receiving magnet, each coil has about 14,500 centimeters of No. 40 wire, or about 3500 turns. These coils each have an impedance of 21,550Ω at a frequency of 7500 cycles per second. They are used connected in series and have a total of 43,100Ω. The driver magnet was wound with 110 turns of No. 24 wire on each spool, giving 20 ohms impedance at a frequency of 7500 cycles per second. They were connected in parallel giving about 10Ω impedance.

The complete set-up for measuring the "specific damping capacity" is shown in Fig. 4. The driving and receiving magnets and the rod support mounted on the brass base are shown at 1. This assembly is mounted on a base 2, the purpose of which is to absorb the
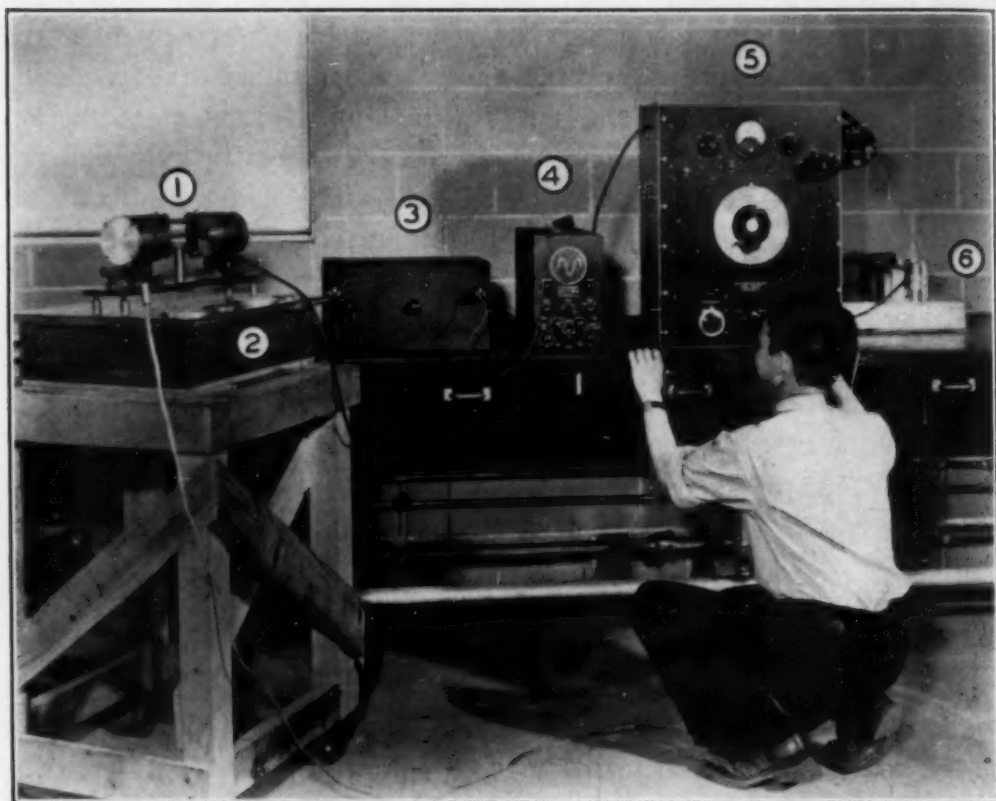
Fig. 4—Complete Assembly for Measuring Vibration Damping Capacity.

vibrations of the floor. It consists essentially of two lead plates weighing about 40 pounds each, with a sheet of sponge rubber between them, the lower plate resting on rubber balls. The whole is enclosed in a wooden box, the top of which is loose and rests upon the upper lead plate.

The power to drive the rod is supplied to the driving magnet from an amplifier 6, which in turn connects to a beat-note oscillator 5, which can supply up to 150 volts at frequencies from 20 to 40,000 cycles per second. The amplifier consists of a 76-driver stage, transformer coupled to a 6L6 push-pull output stage, which connects to the driving magnet through an output transformer having a 10-ohm secondary.

The voltage induced in the receiving magnet by the rod vibrations

is quite small, being in the order of 1 millivolt. This output is therefore connected to an amplifier 3, the output of which connects to a cathode ray oscillograph 4. The amplifier consists of three stages, giving a voltage gain of from 40,000 to 50,000 for frequencies from 1000 to over 40,000 cycles per second. The oscillograph is a conventional commercial product having a sensitivity of 2 volts per inch.

This apparatus can be used for measuring the damping in all kinds of materials in rod form. If the material is magnetic it can be used as it is, the coupling between it and the magnets being enough to obtain effective driving. If it is nonmagnetic but conducting, the eddy currents induced in it will produce adequate coupling to drive the rod fairly well, but in this case it is much better to fasten small sheet iron pieces on the ends of the rod and obtain good magnetic coupling this way. Sealing wax will do to fasten the pieces on the rod ends. If the material is neither magnetic or conducting, the sheet iron end pieces are essential.

STABILITY OF VIBRATION DAMPING CAPACITY

Preliminary observations indicated that quenched alloys showing very high initial vibration damping capacities lost this vibration damping capacity to some extent with time. It was therefore necessary to determine the conditions under which stable vibration damping capacities could be obtained. These results are shown in Table I.

Table I
Stability of Vibration Damping Capacity

Rod	Composition, %			Heat Treatment		Vibration Damping Capacity on Standing				
	Mn	Cu	Ni	Quench	Temp. °C.	0 Hr.	48 Hr.	72 Hr.	120 Hr.	264 Hr.
K-26	90.3	9.8	...	Oil	1025	11.1	5.2	...	3.2	....
K-29	82.4	17.8	...	Oil	950	....	8.8	...	6.1	....
H-89	90.5	8.4	...	...	1050	2.18	0.84	...	0.51	....
K-26	90.3	9.8	...	H_2O	500	....	3.7	...	3.7	....
K-42	65.0	30.0	5.0	H_2O	450	1.75	...	...	...	1.71
K-29	82.4	17.8	...	H_2O	500	3.4	...	4.1	...	....
K-31	49.7	50.6	...	H_2O	500	3.0	...	3.6	...	....

From this we conclude that alloys reheated to temperatures from 450 to 600 degrees Cent. are stable with regard to vibration damping capacity.

EFFECT OF COLD WORK ON VIBRATION DAMPING CAPACITY

In a previous publication we have noted that cold work greatly reduces vibration damping capacity of high manganese-copper alloys.

The reduction in vibration damping capacity over that of stable quenched alloys is somewhat less than was indicated in that publication. There is, however, a definite reduction in the alloys with more than 75 per cent manganese. The vibration damping capacity of alloys with less than 75 per cent manganese is relatively low in the quenched state and is not measureably lowered by cold work.

Effect of Heat Treatment

The effect of heat treatment on the vibration damping capacity of the alloys containing more than 40 per cent manganese is very
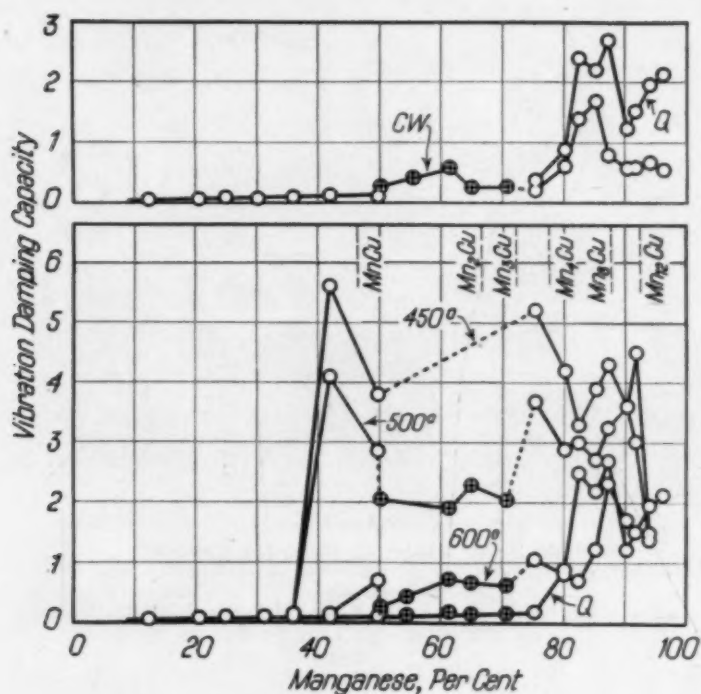
Fig. 5—Vibration Damping Capacity of Manganese-Copper Alloys After Cold Working Quenching and Reheating to Various Temperatures.

great. Referring to Table II, it is seen that by heat treatment at 450 degrees Cent. the maximum vibration damping capacity is obtained in all but the highest manganese alloys. The values in Table II are shown graphically in Fig. 5. The limit of stable solid solution at approximately 40 per cent manganese is revealed much more sharply by vibration damping capacity than by the electrical resistance previously considered. In general, however, the change of vibration damping capacity parallels the change of resistance up to 75 per cent

manganese. Above 75 per cent manganese the vibration damping capacity is less affected by heat treatment, the resistance much more so. It should also be pointed out that whereas electrical resistance passes through a minimum at 500 degrees Cent. for all but the highest manganese alloys, vibration damping capacity passes through a maximum at a definitely lower temperature around 450 degrees Cent.

Table II
Alloy Number, Composition and Vibration Damping Capacity After Cold Working, Quenching and Reheating to Various Temperatures

| Rod | Composition | | Cold- | Quenched | Vibration Damping Capacity | | | | | |
	Mn	Cu	Worked		350° Q	400° Q	450° Q	500° Q	600° Q	700° Q
K-15	50.1	49.8	0.28	0.19	....	....	....	2.04	0.28	0.10
16	55.4	44.4	0.42	0.13	....	....	....	2.36	0.44	0.17
17	61.3	38.7	0.59	0.18	....	....	....	1.92	0.73	0.16
18	64.9	34.7	0.17	0.15	....	....	....	2.3	0.67	0.11
19	70.7	29.5	0.18	0.16	....	....	....	2.05	0.63	0.13
23	96.2	3.9	0.55	2.13	....	....	....	....	....	....
24	94.0	6.1	0.67	1.96	1.45	1.43	1.51	1.39	....	....
25	91.7	8.0	0.59	1.52	2.68	2.17	4.46	3.02	....	....
26	90.3	9.8	0.58	1.23	2.10	1.91	3.55	2.61	1.72	1.33
27	87.2	12.5	0.79	2.7	3.51	3.45	4.26	3.25	2.32	....
28	85.2	14.9	1.68	2.22	2.75	2.27	3.91	2.72	1.23	1.12
29	82.4	17.8	1.39	2.46	4.14	3.12	3.26	3.00	0.71	0.61
30	80.1	20.0	0.61	0.88	1.22	3.96	4.15	2.89	0.82	0.44
31	75.3	23.8	0.21	0.19	0.21	4.54	5.22	3.69	1.06	0.14
32	49.7	50.6	0.11	0.10	0.10	0.19	3.76	2.86	0.70	0.10
33	41.9	57.8	0.13	0.09	0.08	0.09	5.56	4.10	0.14	0.09
34	35.9	64.4	0.10	0.09	....	0.09	0.11	0.14	0.10	0.09
35	31.1	69.3	0.07	0.09	....	0.09	0.07	0.07	0.09	0.08
36	24.9	75.5	0.09	0.09	....	0.08	0.08	0.08	0.08	0.09
37	20.5	79.2	0.06	0.07	....	0.07	0.06	0.08	0.08	0.08
38	12.3	87.8	0.05	0.05	....	0.06	0.06	0.07	0.08	0.06

These considerations lead us to suggest that whereas the change of electrical resistance is associated with two processes which are co-active over certain temperature ranges, vibration damping capacity is increased to any extent solely by the lower temperature process. This process is in all probability an ordering of the manganese-copper lattice. The second process, which is accompanied by separation of a phase, evidently lowers vibration damping capacity directly or indirectly.

Consideration of the results in Table III indicates that the lower temperature at which maximum vibration damping capacity is obtained as compared with minimum resistance probably is due to the fact that vibration damping capacity reaches its maximum with the presence of a large number of anti-phase nuclei in the alloy, rather than when the alloy has been brought into substantially equilibrium order. Thus we find that the slow cooled alloy K-29a has a vibration damping capacity of only 1.88, whereas the quenched and reheated samples have considerably higher vibration damping capacities.

In the case of alloy K-33a, we find that the slow-cooled alloy has a slightly higher vibration damping capacity than the quenched alloy but has a very much lower value than the alloy, which is

Table III

Alloy Number, Composition and Vibration Damping Capacity After Cold-Working, Reheating to 450 Degrees Cent., Quenching, Reheating to 450 Degrees Cent., and Slow Cooling in the Furnace from the Quenching Temperature

| Rod | Composition | | Cold-Worked | Vibration Damping Capacity | | | | |
	Mn	Cu		Reheat to 450° After Cold-Working	Quenched	Reheat to 450°	Slow Cooled
K-29a	81.2	18.9	0.48	3.51	4.36	3.29	1.88
K-33a	39.5	60.4	0.08	0.37	0.10	1.88	0.23

quenched and reheated to 450 degrees Cent. The inability to obtain more than qualitative checks of the vibration damping capacity of these alloys by repeating given treatments which are indicated, for example, by comparison between the original alloys K-29 and K-33 and other melts of quite similar composition represented by K-29a and K-33a, is in our opinion due to the extreme sensitiveness of vibration damping capacity to past treatment, such as rate of heating and cooling and other conditions that could be duplicated only by special care.

The effect of intermediate cold work between the quenching and reheating is very pronounced in the lower-manganese alloys, such as K-33a. In this instance, reheating the cold-worked alloy to 450 degrees gave a vibration damping capacity of only 0.37, while reheating after quenching gave a figure of 1.88 which, although relatively high, is still very much below the value of 5.56 obtained for the original rod of K-33.

It should be pointed out that there is a definite difference in the treatments given rod K-33 and K-33a. K-33 was heated for 2 hours at a series of increasing temperatures, as indicated in the table, while rod K-33a was heated only at 450 degrees Cent. for 2 hours.

EFFECT OF STRESS BELOW THE YIELD POINT ON VIBRATION DAMPING CAPACITY

Since vibration damping capacity is an expression of elastic hysteresis, the application and removal of a stress might be expected to affect the vibration damping capacity at least for a considerable period until the dimensions of the lattice returned to normal. Our

investigations showed this to be the case. Rods were placed in a tensile machine and a load applied. The load was then removed and the set noted on an extensometer. After each such treatment, the vibration damping capacity of the rods was measured. The results are shown in Table IV. The specific example chosen to illustrate this effect is actually a ternary alloy, but less complete data on the binary alloys show similar trends. Three thousand pounds is about the elastic limit of this specimen, and the cold working produced by higher stresses causes the vibration damping capacity to decrease.

Table IV
Effect of Stress on Vibration Damping Capacity

Rod	Composition Mn	Cu	Ni	Treatment	Vibration Damping Capacity at Various Times 0 Hr.	18 Hr.	24 Hr.	264 Hr.
K-42	65	30	5	Cold-Worked	0.103	...	...	....
				450° C.—Quench H_2O	1.75	...	...	1.71

Elapsed Time Hr.	Total Tensile Stress, in Pounds 500	1000	1500	2500	3000	3500	4000
0	1.89	1.95	1.92	2.09	1.95	1.73	1.3
18	....	....	....	1.79	....	....	0.97

SUMMARY

The vibration damping capacity at low stresses has been measured for a series of manganese-copper alloys. The results taken, together with those previously published on electrical resistance and microstructure which are given in other papers of this series, indicate that the high vibration damping capacity obtained in this alloy system by certain heat treatments is due to the formation of ordered antiphase nuclei in the mass of the alloy.

The vibration damping capacity reaches a maximum with some critical distribution of these anti-phase nuclei and decrease as a state of equilibrium order is approached. The effect of cold work is to decrease the vibration damping capacity of the alloys, as would be expected from its disordering effect. The unusually high vibration damping capacity of the quenched alloys would seem at first glance to vitiate the conclusion that the high vibration damping capacity is due to order. However, it is shown that the high vibration damping capacity of quenched alloys drops with time, and it is suggested that the high vibration damping capacity in the quenched alloys is due to their failure to come to an equilibrium volume when quenched.

There would thus seem to be a similarity in structure with regard to its effect on vibration damping capacity between an alloy in a supercooled state and an alloy in a partly ordered state. In both instances, some of the atoms of the metal are not in the equilibrium positions required by the temperature. High vibration damping capacity appears to be an attribute of such metastable systems.

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DISCUSSION

Written Discussion: By R. L. Wegel, Bell Telephone Laboratories, New York.

Mr. Dean and his associates are to be congratulated on their discovery of a metal having the very unusual property of high internal damping associated with relatively high elasticity and strength. This manganese-copper alloy, when prepared as the authors have shown, has a damping capacity very similar to that of lead and other mechanical properties comparable with brass. It is to be expected that this peculiar combination of properties will be found technically useful.

The change in damping capacity with cold work and heat treatment, no doubt results from a complex of causes. The reasoning by means of which the authors conclude that the ordering of atoms in the lattice is a large contributor is not entirely clear. After the appearance of the first paper by the authors (November, 1939) a small number of tests of this material was made at the Bell Telephone Laboratories by Mr. H. Christensen with results which fit in very well with those of the present paper. On three samples, heat treated in different ways, he made measurements of the variation of hardness from point to point on their surfaces. The results seem to indicate that the highest damping was correlated with the largest variation in hardness, that is, with the greatest gross in homogeneity.

Authors' Reply

Dr. Wegel's suggestion that the high vibration damping capacity of these alloys might be correlated within homogeneities of such large dimensions that they could be detected by hardness measurements is interesting. We find it hard to think of the damping phenomenon as concerned with such major inhomogeneities. We have not noticed such hardness variations in our alloys and after Dr. Wegel's suggestion, we checked many of the alloys which we had measured and were unable to find any hardness variation beyond the experimental error.

EQUILIBRIUM RELATIONS IN THE SOLID STATE OF THE IRON-COBALT SYSTEM

By W. C. Ellis and E. S. Greiner

Abstract

There are important transformations in the solid in the iron-cobalt system. One of these originates from the A_3 transformation in iron. Cobalt in the binary system at first raises the A_3 transformation to a maximum in the region of 45 weight per cent cobalt. Further additions decrease the temperature of transformation which rapidly approaches room temperature in the region of 80 weight per cent cobalt. An extended two phase region from 76.5 to 88.5 weight per cent cobalt was established at 600 degrees Cent. (1110 degrees Fahr.).

An order-disorder transition occurs in the alpha phase in the region of 50 weight per cent cobalt. The critical temperature of order is in the neighborhood of 700 degrees Cent. (1290 degrees Fahr.) depending upon the composition. The ordered arrangement has the cesium chloride structure.

The lattice constants of the alpha phase deviate widely from a linear function of the cobalt content. The first additions of cobalt increase the cell size to a maximum at approximately 20 per cent cobalt. Further additions result in a contraction in the cell size to the limit of the alpha phase. Compositions in the region of 50 per cent cobalt exhibit an increase in cell size on ordering.

EQUILIBRIUM relations in the solid of the iron-cobalt system are of particular interest because of the increasing importance of iron-cobalt alloys as magnetic materials, and because of the rapid expansion in the number of multi-component alloys in the fields of magnetic materials, tool compositions, and materials of special properties in which iron and cobalt are major constituents. Features of the iron-cobalt system have been investigated by Guertler and Tammann (1),[1] Ruer and Kaneko (2), Kase (3), Preuss (4), Andrews (5), Ellis (6), Osawa (7), Masumoto (8), Hashimoto (9), Andrews and Nicholson (10), and Rodgers and Maddocks (11). The

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[1]The figures appearing in parentheses refer to the bibliography appended to this paper.

work of these investigators with the exception of the last two references are critically reviewed in "Der Aufbau der Zweisstofflegierungen" by M. Hansen. References (10) and (11) deal with recent investigations of the equilibrium in the solid state and together with Masumoto's important work (8) on the same subject will be discussed in connection with the current investigation.

The researches of Masumoto (8) and Andrews and Nicholson (10) have located the alpha-gamma phase transformations in iron-cobalt alloys. The reported data are plotted in Fig. 2. There is disagreement in the magnitude and location of the two phase area in the region of 80 weight per cent cobalt. The latter investigators found by microscopic methods a two phase region extending from 74 to 81 weight per cent cobalt at 500 degrees Cent. (930 degrees Fahr.) contrary to the narrow area at 78 weight per cent cobalt established by Masumoto by dilatometric methods.

Rodgers and Maddocks (11) working with the same alloys as were used by Andrews and Nicholson (10) determined the lattice constants of the alpha phase alloys (0-76 weight per cent Co.) after one heat treatment, heating at 600 to 650 degrees Cent. (1110 to 1200 degrees Fahr.) for 2 hours in hydrogen and slowly cooling. Significant features were disclosed. The first additions of cobalt to iron result in a small increase in cell size reaching a maximum at about 20 weight per cent cobalt. Further increases in cobalt content result in a decrease in cell size. In the region at 45 per cent cobalt there is an upward trend in the curve developing into a cusp at 48 weight per cent. They further established by X-ray methods through the appearance of extra reflection lines that, in the alloy corresponding to a composition near to FeCo, an ordered arrangement of the cobalt and iron atoms exists after the heat treatment used. An order-disorder transition had been long suspected in this region in the iron-cobalt system because of the anomalous nature of the electrical conductivity-composition curve, and was predicted by Kussmann, Scharnow and Schulze (12) in 1932.

SCOPE OF PRESENT INVESTIGATION

In the present investigation the alpha-gamma phase transformation and the Curie points for low cobalt compositions have been redetermined. For the region to 70 per cent* of cobalt a thermal

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*Compositions unless otherwise specified are in atomic per cent.

method was used. In the region of 80 per cent cobalt, where the phase boundaries are nearly vertical the limits of the two phase area were established by X-ray diffraction methods.

The lattice constants for the alloys in the alpha phase were determined after bringing to equilibrium at 800 degrees Cent. (1472 degrees Fahr.) and 575 degrees Cent. (1067 degrees Fahr.) followed by quenching. As will be discussed, significant differences were obtained in lattice constant values after these two treatments for alloys in the region of 50 per cent cobalt. The existence of the superlattice was confirmed in alloys in the region of 50 per cent cobalt, the composition limits for superstructure existence were established, and the temperature-composition curve of the critical temperature of order was determined.

METHODS OF INVESTIGATION

Preparation of Materials—Iron and cobalt of electrolytic origin were used for preparing the alloys. The iron contained 0.02 weight per cent carbon, 0.008 weight per cent phosphorus, and 0.002 weight per cent sulphur. The cobalt contained 0.07 weight per cent iron and 0.02 weight per cent copper as principal impurities. The sulphur content was less than 0.001 weight per cent. Spectroscopic tests revealed no appreciable amounts of additional impurities in either the iron or cobalt. Ingots of 100 to 200 grams were prepared by fusion and solidification in alundum crucibles in vacuum. The ingots were homogenized by heating at 1000 degrees Cent. (1830 degrees Fahr.) for approximately 168 hours alternately in hydrogen and in vacuum; the heating was finished in vacuum. Samples were taken from the homogenized ingots for chemical analyses, for X-ray diffraction measurements and for thermal analyses. The analyzed compositions of the homogenized ingots are given in Table I.

Method of Thermal Analysis—Phase transformation temperatures in compositions containing to 70 per cent cobalt, Curie points, and critical temperatures of ordering were determined using the method of thermal analysis described by C. S. Smith (13). A temperature difference across the wall of the refractory container of approximately 60 degrees Cent. (140 degrees Fahr.) was maintained. Temperatures were measured with a calibrated alumel-chromel couple and high sensitivity potentiometer. Time intervals were established with a split second, double action stop watch.

Table I
Composition, Heat Treatment and Structure of Alloys

Sample No.	Composition by Analysis — Iron Wt. Per Cent	Cobalt Wt. Per Cent (By Diff.)	Iron At. Per Cent	Cobalt At. Per Cent (By Diff.)	Heat Treatment Quenching Temperature, Deg. C.	Time at Quenching Temperature, Hours	Phases Observed	Lattice Constant-Angströms at 25 Deg. C. (77 Deg. F.) Alpha Phase	Gamma Phase
1-A-1	100.0	0.0	100.0	0.0	575	740	Alpha	2.8607	
1-A-2					600	5	Alpha	2.8607	
2-A-1	88.74	11.26	89.27	10.73	575	740	Alpha	2.8611	
2-B-2					800	6	Alpha	2.8610	
3-A-1	78.63	21.37	79.52	20.48	575	740	Alpha	2.8610	
3-B-2					800	6	Alpha	2.8611	
26-A-1	73.33	26.67	74.37	25.63	575	920	Alpha	2.8603	
26-B-1					800	6	Alpha	2.8602	
4-A-1	68.41	31.59	69.56	30.44	575	740	Alpha	2.8583	
4-B-1					800	5	Alpha	2.8585	
5-A-1	58.20	41.80	59.51	40.49	575	740	Alpha	2.8542	
5-B-1					800	5	Alpha	2.8541	
23-A-1	53.62	46.38	54.96	45.04	575	920	Alpha	2.8529	
23-B-1					800	5	Alpha	2.8517	
27-A-1	51.86	48.14	53.21	46.79	575	240	Alpha	2.8523	
27-A-2					575	1100	Alpha	2.8522	
27-B-1					800	5	Alpha	2.8511	
24-A-1	50.22	49.78	51.57	48.43	575	920	Alpha	2.8515	
24-B-1					800	5	Alpha	2.8502	
6-A-1	48.35	51.65	49.70	50.30	575	740	Alpha	2.8504	
6-A-2					575	670	Alpha	2.8504	
6-A-4					800*	2	Alpha	2.8508	
6-A-5					720**	64	Alpha	2.8513	
6-B-1					800	5	Alpha	2.8486	
6-B-2					800	5	Alpha	2.8490	
25-A-1	43.56	56.44	44.89	55.11	575	920	Alpha	2.8474	
25-B-1					800	6	Alpha	2.8463	
7-A-1	38.38	61.62	39.66	60.34	575	740	Alpha	2.8435	
7-B-1					800	5	Alpha	2.8426	
8-A-1	28.77	71.23	29.89	70.11	575	740	Alpha	2.8370	
8-B-1					800	5	Alpha	2.8369	
20-A-1	25.20	74.80	26.23	73.77	575	1000	Alpha	2.8345	
9-A-1	19.19	80.81	20.04	79.96	580	670	Alpha + gamma	2.8337	3.5490
9-B-1					800	5	Alpha + gamma	2.8348	3.5566
9-D-1					680	110	Alpha + gamma	2.8338	3.5516
19-A-1	14.12	85.88	14.79	85.21	575	1000	Alpha + gamma	2.8337	3.5484
19-B-1					800	5	Gamma		3.5529
19-C-1					965	3	Gamma		3.5529
10-A-1	9.69	90.31	10.17	89.83	580	670	Gamma		3.5471
10-A-1					580	670	Gamma		3.5473
10-B-1					800	5	Gamma		3.5474
18-A-1	4.66	95.34	4.91	95.09	575	1000	Gamma		3.5419
18-B-1					800	5	Gamma		3.5419

*Slow cooled (<3° C./min.)
**Slow cooled, 15° C./day, from 720° to 400° C.

Methods of X-ray Diffraction Analysis—The samples for X-ray diffraction analysis were prepared by collecting the powders obtained by grinding the homogenized ingots on an alundum wheel. The metallic powder was separated magnetically from the alundum powder. The metal powders were sealed in evacuated quartz or Pyrex glass tubes and heat treated as noted in Table I. The specimens for treatment below 600 degrees Cent. (1110 degrees Fahr.) were heated in a horizontal furnace and quenched by quickly removing the tubes from the furnace and plunging into water. The specimens treated at

higher temperatures were quenched extremely quickly by means of a quenching technique, which has been described by Jette (14) and his co-workers.

The lattice constants of the heat treated materials were determined from X-ray reflection patterns obtained in a back-reflection focusing camera described by Jette and Foote (15). Reflection angles were corrected (15) for refraction. Lattice constants were calculated by the method of Cohen (16) which eliminates the systematic errors inherent in this type of reflection pattern. The precision of the lattice constants was approximately one part in thirty thousand.

In order to obtain satisfactory distributions of reflections, two element targets were used in the X-ray tube. The lattice constants of the body-centered cubic or alpha phase alloys were determined from Fe-Kα (220) and Co-Kα (310) reflections, while the lattice constants of the face-centered cubic or gamma phase alloys were determined from Fe-Kα (222) and Mn-Kα (311) reflections. The X-ray wave lengths were those assigned by Siegbahn (1931).

For detecting the presence of an ordered arrangement in the alloys near to the composition, FeCo, a different procedure was required. Cobalt-Kα_1 radiation was used since it enhances the normal small difference in scattering power of iron and cobalt atoms, and would make possible from theoretical considerations (17) the superlattice reflections when ordering exists. The superlattice reflections were obtained in a focusing camera of the type for recording the low, medium and high angle reflections.

The Alpha-Gamma Phase Transformation

Compositions From 0-70 Per Cent Cobalt—The temperatures of phase change in the compositions from iron to 70 per cent cobalt, the Curie points in the low cobalt compositions, and the critical temperatures of order were determined by the thermal method (13). Thermal data were obtained for both heating and cooling. The temperature spread in the thermal arrests for the alloys containing to, and including, 50.3 per cent cobalt were small. For these compositions the temperature of the maximum effect, the cusp in the inverse rate curve, was taken as a mean temperature for the transformation and is recorded in Table II for both heating and cooling. The cusp temperatures on heating and cooling were averaged to give the equilibrium transformation temperatures. These points have been

Table II
Thermal Analysis Data

Alloy No.	Cobalt Content Wt. %	At. %	Temperature of Phase Change Deg. C. Heating	Cooling	Mean Value	Temperature of Magnetic Change Deg. C. Heating	Cooling	Mean Value	Critical Temperature of Order Deg. C. Heating	Cooling	Mean Value
1	0.0	0.0	918*	900*	909	770	766	768			
2	11.26	10.73	928*	918*	923	884	878	881			
3	21.37	20.48	950*	946*	948						
4	31.59	30.44	976*	970*	973						
5	41.80	40.49	987*	984*	986			—	712	713	713
23	46.38	45.04							735	729	732
6	51.65	50.30	983*	974*	979				732	728	730
25	56.44	55.11							716	716	716
7	61.62	60.34	958**	942***	950				680	676	678
			962***	945**	954						
8	71.23	70.11	908**	862***	885						
			915***	869**	892						

*Temperature of longest period of thermal arrest.
**Temperature of beginning of transformation.
***Temperature of end of transformation.

denoted by the hollow circles in Fig. 1. In addition, a more detailed analysis of the thermal curves was made to locate the beginning and end of the transformation for compositions in this region. The method was the same as used for the compositions containing higher cobalt percentages described in the next succeeding paragraph. The analysis resulted only in an approximation, in that there is, in general, a rounding off of the thermal curves rather than a sharp break at the onset and completion of the transformation. The approximations are considered, however, to have established the phase boundaries within a few degrees in that the temperature ranges involved are small. The results are shown by the two full lines on either side of the "hollow circle" points in Fig. 1.

In the compositions containing 60.34 and 70.11 per cent cobalt there was a substantial temperature difference between the beginning and end of the transformation. In these instances the initial and final temperatures of transformation on heating and cooling were respectively averaged and the two points thus obtained plotted in Fig. 1. These two points establish the boundaries of the two phase area for each composition.

Extended Two Phase Area in Region of 80 Per Cent Cobalt— Since the steepness of the phase boundaries of the two phase region at 80 per cent cobalt renders difficult the location of the phase limits by thermal techniques, an X-ray diffraction method was used. The lattice constants of the alpha and gamma phases in the heat treated two phase alloys were referred to those of the corresponding single

phase alloys. Presumably the two phase alloys were in equilibrium at the respective temperatures and the phases present were retained in the high temperature state, except for the normal thermal contraction.

To clarify the method, portions of the lattice constant-atom per cent cobalt plots for both the alpha and gamma phases are shown in Fig. 3. The lattice constants of these phases in the two phase alloys
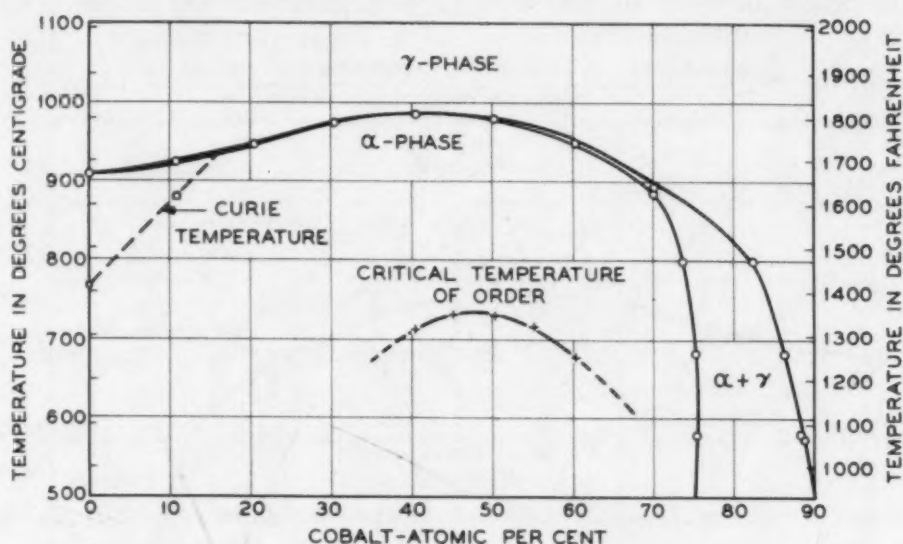
Fig. 1—Equilibrium in the Solid in the Iron-Cobalt System. Points to and Including 70 Per Cent Determined by Thermal Methods; Points in the Extended Two Phase Area by X-Ray Diffraction Methods.

quenched from the temperatures indicated in this figure are designated by the solid circles. The intersections of horizontal lines through these points with the respective lattice constant curves in single phase alloys are the compositions at the solubility limits at the temperature from which the alloy was quenched. The dotted lines connect the compositions of the alpha and gamma phases which are in equilibrium at the respective quenching temperatures. The results are plotted in Fig. 1.

Discussion of Phase Transformation—The temperatures obtained in the present investigation together with those of Masumoto (8) and of Andrews and Nicholson (10) for the alpha-gamma transformation are plotted in Fig. 2. In the absence of specific experimental points or table of data the points of Andrews and Nicholson were taken from their published curve. The location of the transformation in the present study lies, for the most part, for cobalt contents to 62 per cent at temperatures between those of the earlier

investigators. An important exception is in the region of low-cobalt compositions where Andrews and Nicholson (10) obtained an initial lowering of the transformation temperature based largely on a value of A_3 for iron substantially higher than the generally accepted value of 910 ± 1 degree Cent. (1670 ± 2 degrees Fahr.).

Disagreement among investigators exists in the location and extent of the two phase area in the region of 80 per cent cobalt. The dilatometric method of Masumoto (8) gave at 600 degrees Cent.
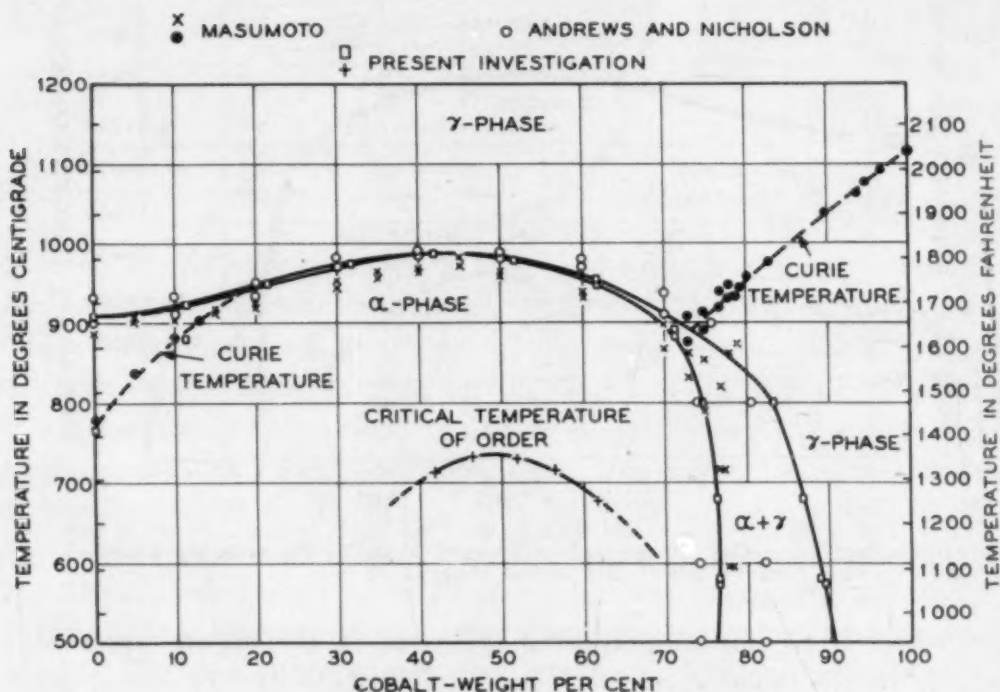
Fig. 2—Equilibrium in the Solid in Iron-Cobalt System.

(1110 degrees Fahr.) only a narrow two phase region extending over 1 or 2 per cent and centered at 78 weight per cent cobalt. Andrews and Nicholson (10) from microscopic studies reported a two phase region at 600 degrees Cent. (1110 degrees Fahr.) extending from 74 to 81 weight per cent cobalt. A still wider region was established in the present study by X-ray diffraction methods extending at the temperature of comparison from 76.5 to 88.5 weight per cent cobalt. The validity of the larger area which was established by quantitative lattice constant measurement is confirmed qualitatively by the presence of reflections originating from both the alpha and gamma phases in the alloy containing 80.81 weight per cent cobalt when quenched from 800 degrees Cent. (1470 degrees Fahr.), and in the

alloy containing 85.88 weight per cent cobalt when brought to equilibrium at 575 degrees Cent. (1065 degrees Fahr.) and quenched. The evidence for the more extended two phase area appears entirely sufficient.

The curve drawn in Fig. 2 is presented as the best weighted location for the alpha-gamma transformation in the system iron-cobalt. The addition of cobalt to iron increases the temperature of transformation to a maximum in the region from 40 to 50 weight
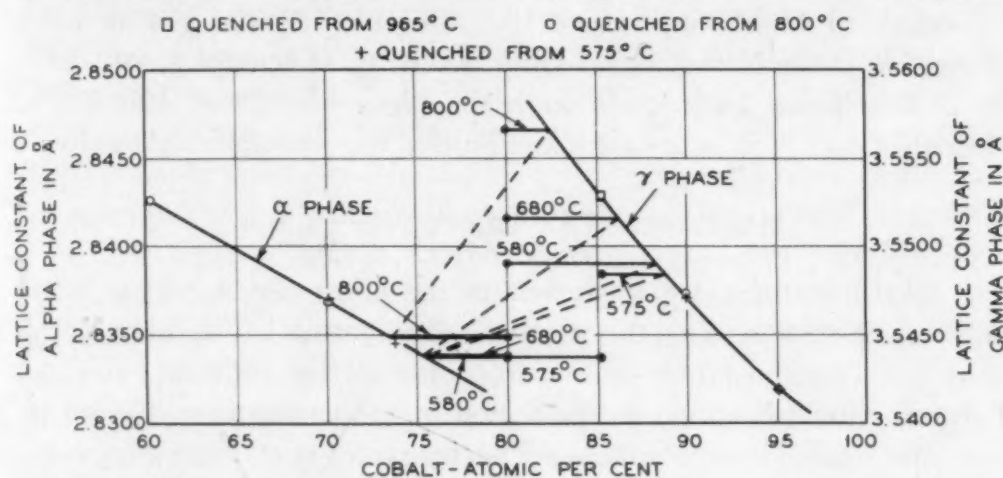
Fig. 3—Determination of Boundaries of Two Phase Region in Iron-Cobalt System.

per cent cobalt. The two phase area is narrow in this range of compositions. From the 50 per cent region the temperature of transformation decreases with further cobalt contents developing into an almost vertical descent in the region of 80 per cent cobalt and a two phase area extending over approximately 13 per cent in composition.

There are two additional features for high cobalt contents, in the iron-cobalt equilibrium in the solid which originate from transformations in pure cobalt. Since the scope of the present study did not extend beyond 95 per cent cobalt these are not shown in Fig. 2 and will be but briefly discussed here. One of these transformations which occurs at a temperature, 400 to 450 degrees Cent. (752 to 840 degrees Fahr.), has been definitely established (6), (18) by thermal, electrical conductivity, and X-ray diffraction measurements, to be a change on heating from a close-packed hexagonal to a face-centered cubic arrangement. The face-centered cubic phase is identical with the gamma phase shown in Fig. 2. The addition of iron to cobalt rapidly lowers the temperature of this transformation.

The other transformation occurring at a higher temperature

should probably be considered at this time as suggested rather than as definitely established. There is disagreement among investigators (18) as to the existence of this phase transformation, and in addition among those reporting a high temperature modification, a divergence of opinion as to the temperature. The temperature has been variously fixed at 850 degrees Cent. (1562 degrees Fahr.) and approximately 1020 degrees Cent. (1870 degrees Fahr.) by investigators using different methods. An investigation of the crystal structure of cobalt at high temperatures by Marick (18) did not reveal a change in phase in the region from 800 to 1200 degrees Cent. (1470 to 2190 degrees Fahr.), although the phase change at 400 to 450 degrees Cent. (750 to 840 degrees Fahr.) was conclusively disclosed.

Phenomena in the Alpha Phase

Three features are important in the alpha phase region in the iron-cobalt system: (a) the magnetic transformation for low cobalt contents, (b) the effect of cobalt on the lattice constant, and (c) "ordering" which occurs in the region of the composition FeCo.

Magnetic Transformation—The magnetic transformation curve (Curie point curve) in the alpha phase is shown in Fig. 1 as the dotted curve extending from 765 degrees Cent. (1410 degrees Fahr.) for iron to the alpha-gamma transformation at 15.5 per cent cobalt. The transformation temperature was taken as the temperature at which the maximum heat effect occurred in the thermal analysis. The data are in satisfactory agreement with the earlier curve of Masumoto (8) as shown in Fig. 2.

Lattice Constants of the Alpha Phase—The lattice constants of the alloys of the alpha phase were determined at 25 degrees Cent. (75 degrees Fahr.) for equilibrium at 800 degrees Cent. (1470 degrees Fahr.) and at 575 degrees Cent. (1065 degrees Fahr.). The change in lattice constant with cobalt content is shown in Fig. 4 by the full and dotted lines. The lattice constants of the compositions heat-treated at 800 degrees Cent. (1470 degrees Fahr.) and at 575 degrees Cent. (1065 degrees Fahr.) increase from 2.8607 to 2.8611 Å as the composition changes from 0 to 20.5 per cent cobalt. Further increases in cobalt content result in a decrease in the size of unit cell, but to approximately 40 per cent cobalt the cell constant values are the same for the two heat treatments within the small experimental error.

Between 40 and 67 per cent cobalt an extremely significant feature occurs in the lattice constant curve; the cell size is larger after quenching from 575 degrees Cent. (1065 degrees Fahr.) than after quenching from 800 degrees Cent. (1470 degrees Fahr.); the difference becomes a maximum in the region 45 to 50 per cent cobalt and decreases on either side to disappear at the limits cited. The cell size for the composition, 50.3 per cent cobalt, is further increased over that for the 575 degrees Cent. (1065 degrees Fahr.) treatment by slow cooling from 800 degrees Cent. (1470 degrees Fahr.). The maximum effect is secured by extremely slow cooling, cooling from 720 to 400 degrees Cent. (1330 to 750 degrees Fahr.) at the rate of 15 degrees Cent. (60 degrees Fahr.) per day. The data are shown by the single points in Fig. 3. From 67 per cent cobalt to the limit of the alpha phase heat treatment is without effect on cell size.

The lattice constant data obtained by Rodgers and Maddocks (11) have been plotted as the triangles in Fig. 4. The triangles refer to specimens which had been slow-cooled from 600 to 650 degrees Cent. (1110 to 1200 degrees Fahr.). Since it has been shown that the lattice constants of the iron-cobalt alloys containing from 0 to approximately 40 per cent cobalt are not affected by heat treatment, the data obtained in the two investigations on these alloys should be comparable. Reference to Fig. 4 shows that the values obtained by Rodgers and Maddocks (11) for compositions in this region run generally parallel but approximately 0.0005Å lower than those obtained in the present investigation. This difference is probably due to the difference in iron used in preparing the alloys, since the lattice constant of iron (2.8601 Å) reported by Rodgers and Maddocks (11) is lower than that obtained in the present investigation (2.8607 Å). The latter value is in agreement with that obtained by Jette and Foote (15) on several specimens of high purity iron.

It is noteworthy, however, that the bulge in the lattice constant curve for compositions on this region observed for the 575 degrees Cent. (1065 degrees Fahr.) treatment was also obtained by the earlier investigators (11) after a 600 to 650 degrees Cent. (1110 to 1200 degrees Fahr.) treatment. The data are not in agreement in specific details, and undoubtedly this should not be expected in view of the sensitivity in this region of values of lattice constant to heat treatment.

Ordering in Compositions in Region, FeCo—An ordered arrangement of iron and cobalt atoms having the cesium chloride struc-
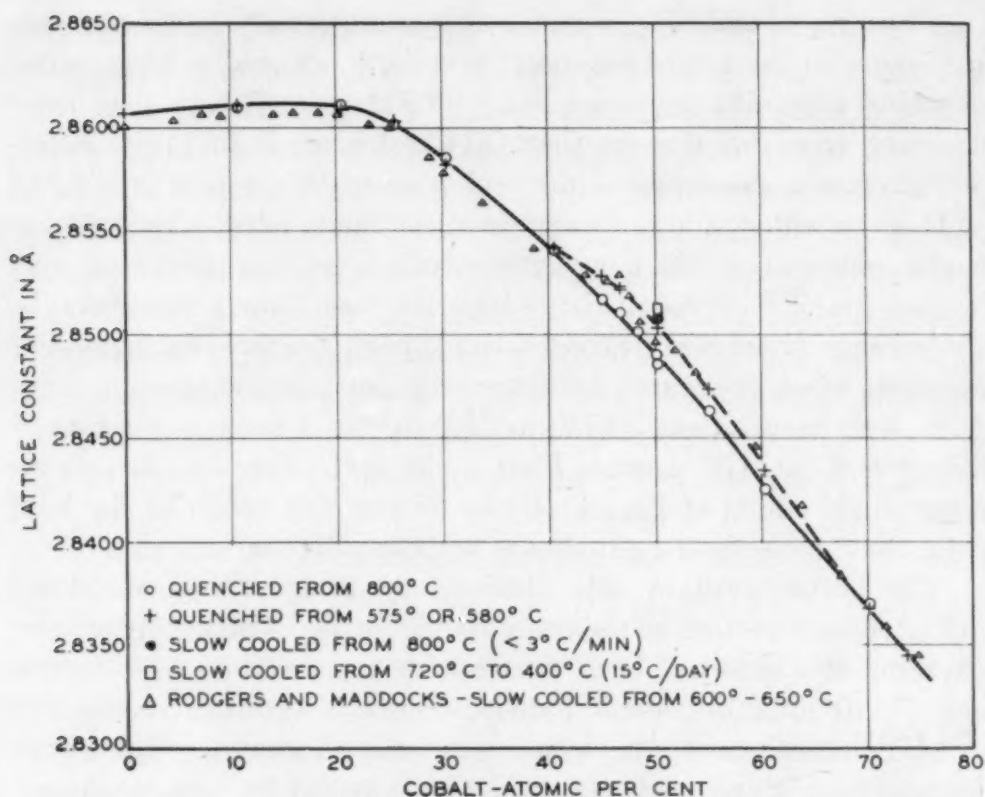
Fig. 4—Lattice Constants of the Alpha-Phase in the Iron-Cobalt System. The Values of the Present Investigation are for 25 Degrees Cent. (77 Degrees Fahr.).

ture was found in the body-centered cubic cell of the composition containing 50.3 per cent cobalt after heating for 720 hours at 575 degrees Cent. (1055 degrees Fahr.) and quenching. This was established by the presence of extra reflections in the X-ray diffraction pattern when $CoKa_1$ radiation was used. The extra reflections which had the (hkl) indices $(1\,0\,0)$, $(1\,1\,1)$, $(2\,1\,0)$ and $(3\,0\,0)$, are indicated in Fig. 5 in relation to the normal reflections for the body-centered cubic lattice. Extra reflections were sought and found also in the compositions containing 40.49, 46.79 and 55.11 per cent cobalt after the above treatment.

No superlattice reflections were observed in the 50.3 per cent alloy heated 5 hours at 800 degrees Cent. (1470 degrees Fahr.) and quenched. This indicates the change in cell size is associated with an order-disorder transition in the lattice of this alloy, the ordered material having a larger cell size than the disordered material. A larger cell on ordering is not usual. It has been reported for the system, iron-palladium (19), in which the volumes of the unit cells increase when the superlattices corresponding to FePd and $FePd_3$ are formed.

Critical Temperature of Order (20)—The presence of an ordered arrangement of cobalt and iron atoms in the alloy containing 50.3 per cent cobalt when at equilibrium at 575 degrees Cent. (1067 degrees Fahr.), together with the absence when equilibrium was established at 800 degrees Cent. (1470 degrees Fahr.) showed that the critical temperature of order for the composition was between
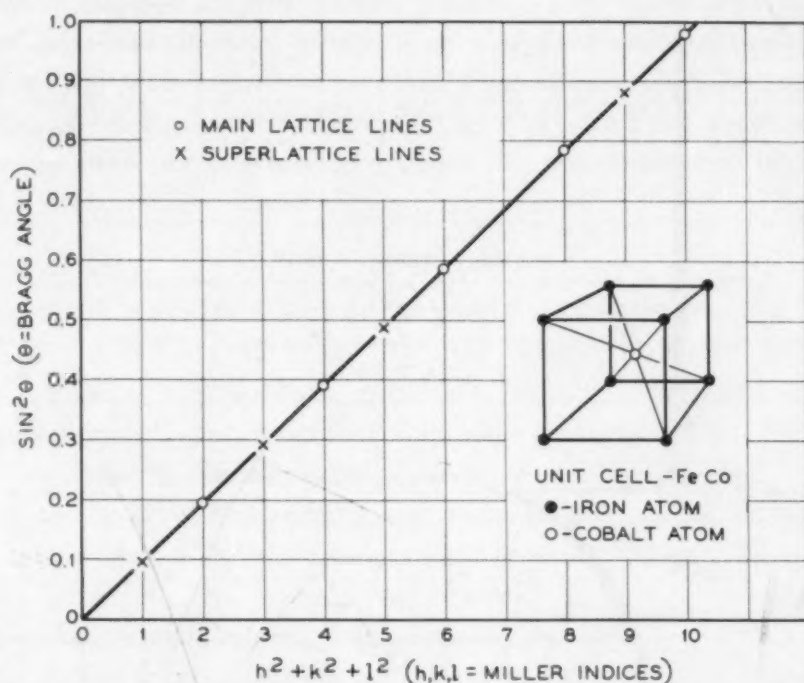
Fig. 5—X-Ray Reflections from Iron-Cobalt Alloy Containing 50.3 Per Cent of Cobalt in Equilibrium at 575 Degrees Cent. (1067 Degrees Fahr.) Using Cobalt-Kα_1, Radiation.

these two temperatures. The temperature was precisely located by thermal analysis for this composition and other compositions in the region of order.

The type of thermal arrest point obtained is illustrated in Fig. 6 for the heating of the composition containing 50.3 per cent cobalt. The curve obtained in this method of thermal analysis is, except at transformation temperatures, an empirical specific heat curve and therefore ideal for investigation of order-disorder transitions. The critical temperature of order for this composition is 732 degrees Cent. (1350 degrees Fahr.) on heating. The thermal arrest temperatures on heating and cooling for the compositions exhibiting ordering are given in Table II and the means, the critical temperatures of order, are plotted in Figs. 1 and 2. The curve of critical temperature
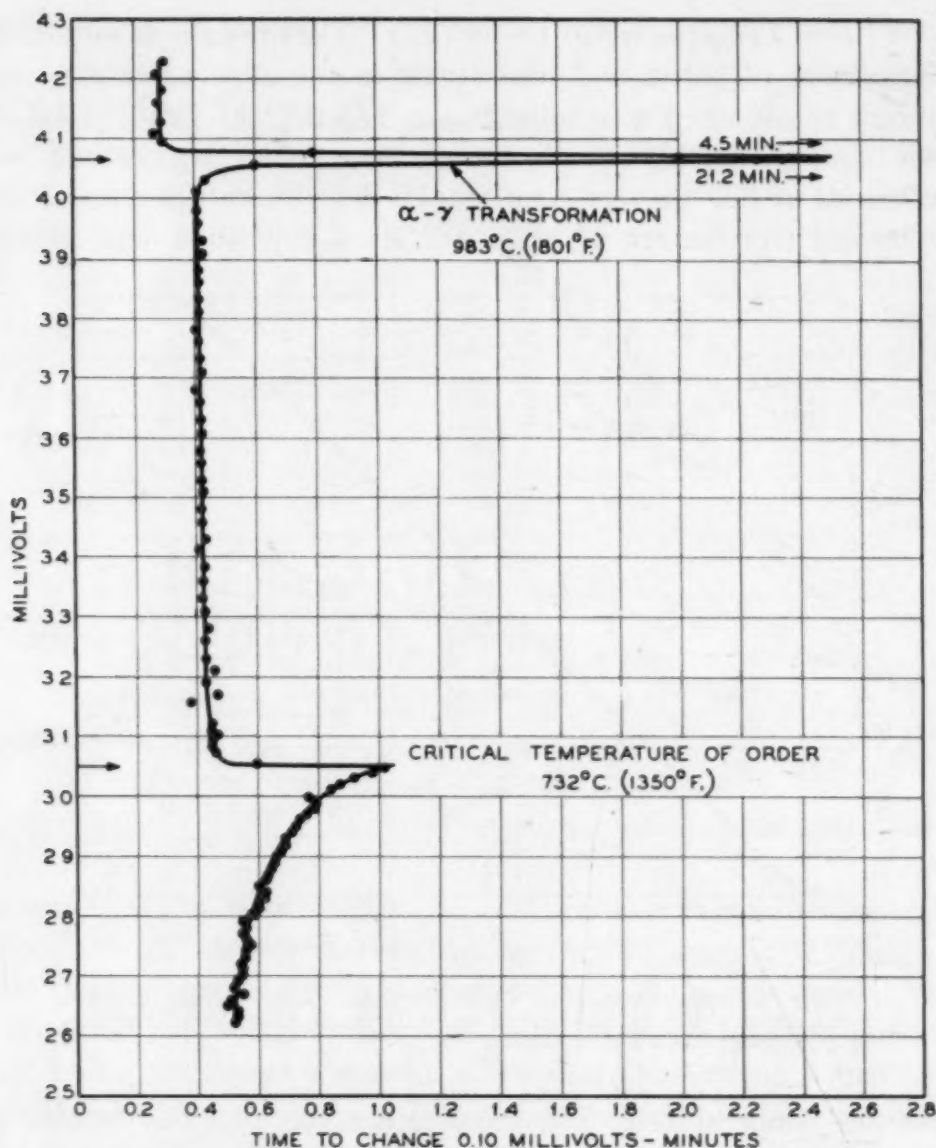
Fig. 6—Heating Curve for the Iron-Cobalt Alloy Containing 50.3 Per Cent Cobalt.

of order has a flat maximum in the region of 45 to 50 per cent cobalt. The precision of the measurements does not admit of a more definite location of the maximum.

In establishing the terminal compositions of the region of order shown in Fig. 1, both thermal effects and cell size change were utilized. The region of order must extend throughout the composition range in which there is a change in cell size after heat treatment below and above the critical temperature of order. One would not be justified however in the assumption that ordering did not occur

outside the composition range exhibiting cell size change. Actually a pronounced "ordering" heat effect was observed in the composition containing 40.4 per cent cobalt although the cell size of this composition was insensitive to heat treatment. The low cobalt limit of the ordering region was set at 35 per cent cobalt since the 30.4 per cent alloy showed no "ordering" thermal effect while the 40.4 per cent composition showed a pronounced effect. The maximum limit at 67.5 per cent cobalt was selected as the first composition whose lattice constant was insensitive to heat treatment taking into account also the data of Rodgers and Maddocks (11) and the experimental fact that the 70.1 per cent cobalt composition exhibited no "ordering" thermal effect.

GENERAL DISCUSSION

The equilibrium relations in the solid in the iron-cobalt system present a number of unusual features which, together with the peculiar property-composition relations, notably electrical conductivity and magnetic permeability, provide a fertile field for theoretical speculation. Earlier investigators have not neglected the possibilities, but have been handicapped by lack of a theory adequate to handle so complex a situation as prevails in this system. Such handicaps still exist and the present authors will be content to point out only the features challenging a consistent explanation.

The alpha solid solutions of iron and cobalt retain stability to higher temperatures with increasing cobalt content to a maximum in the region of 40 to 50 per cent cobalt. Beyond this cobalt content the temperature range of stability of the alpha phase rapidly diminishes until in the region of 80 per cent cobalt the alpha phase no longer exists at room temperature.

The lattice constant curve of the disordered alpha phase exhibits departure from the simple additive rule for solid solutions. The type of deviation, whether positive or negative, depends upon the application of the theory of co-ordination numbers and may be without significance in this system. Experiment has established definitely, however, that the alpha cell increases in size with increase in cobalt contents to 20 per cent cobalt. Beyond this composition, the cell diminishes in size and approximates in the disordered alloys a uniform contraction extending to the limit of the alpha phase. The lattice constant curve shows an upward bulge in the region of the composition of FeCo after "ordering."

Ordering in the region FeCo for the compositions from above 40 to 67 per cent cobalt is accompanied by an increase in cell size. This is anomalous, but has been reported for one other system, iron-palladium (18). It may be significant that both of these systems are composed of transition elements, whose combinations might be expected to give rise to anomalous results because of complexities in the ion structures. In the iron-cobalt system however, the composition at 40 per cent cobalt orders without change in cell size. The critical temperature of order decreases from a maximum in the region of the composition, FeCo. Thus, cell dimension and the critical temperature of order are both composition sensitive.

In the iron-cobalt system, accompanying the described equilibrium features are unusual property-composition relationships. One of these is the high electrical conductivity of alloys in the region of intermediate cobalt contents. In alloys from 20 to 80 per cent cobalt higher conductivities (6) have been obtained than would be expected from the general behavior of solid solutions. This effect not only exists at temperatures where order is present but also persists in measurements at 900 degrees Cent. (1650 degrees Fahr.), which is well above the critical temperature of order. This would indicate that the anomalous electrical behavior cannot be explained wholly as due to ordering. A maximum in magnetic permeability with medium magnetizing forces also occurs in this region. Ordering may be partially responsible but no doubt there are other factors involved in the complex magnetic behavior. A consistent explanation for the observed behaviors in this system awaits both more experimental investigation and further advance in the theory of the metallic state.

SUMMARY

The alpha-gamma transformation in iron-cobalt alloys has been determined using thermal and X-ray diffraction methods. The results have been compared with those of earlier investigators using other methods and a most probable curve is presented. The present investigation shows an extended two phase area in the region at 80 per cent cobalt.

The lattice constants have been determined for the alpha-phase alloys after several heat treatments. This established that the first additions of cobalt increase the cell size. Further additions decrease

the cell constant but in the region of 50 per cent cobalt the cell size is dependent on heat treatment.

The existence of an order-disorder transition in the alpha phase in the region of FeCo has been confirmed. The critical temperatures of ordering for alloys in this region and the composition limits of detectable order are given for the first time. It has been established that ordering results in an increase in the cell size for certain compositions, contrary to the generally observed result for ordering.

It is doubted that the present theories of metallic state are sufficiently complete to explain the detailed behaviors encountered.

In conclusion the authors acknowledge their sincere gratitude to Earle E. Schumacher, under whose supervision this investigation was carried out, and whose continued interest and encouragement greatly aided in the prosecution of the research.

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DISCUSSION

Written Discussion: By W. P. Sykes, metallurgist, Cleveland Wire Works, General Electric Co., Cleveland.

To anyone who has had occasion to ponder over the equilibrium diagram of the iron-cobalt system, this paper is a source of real satisfaction. Earlier findings have been confirmed and new information added by this critical and painstaking investigation and the results can be accepted without question.

I regret that the authors did not extend their investigation to the cobalt end of the series. As they have stated, there is still considerable difference of opinion regarding the existence of a hexagonal form of cobalt at some temperature above 1100 degrees Cent. (2010 degrees Fahr.).

My experience leads me to believe that such is the case, but I have yet to obtain evidence which I should care to put forward as indisputable. The problem is especially difficult because of the ease with which the hexagonal form is produced from the cubic at lower temperature, and its remarkable stability.

The oxide when heated in hydrogen for some 50 hours at 375 to 400 degrees Cent. yields a metal powder which from its diffraction pattern may be identified as hexagonal. Such powder, when heated again at temperatures between 600 and 1000 degrees Cent. (1110 and 1830 degrees Fahr.) in hydrogen for a period of 100 hours will show but one or two weak reflections from the cubic lattice.

On the other hand, the oxide reduced at 600 to 1000 degrees Cent. (1110 to 1830 degrees Fahr.) will yield the cubic pattern frequently accompanied by one or two weak reflections from the hexagonal form. Even such slight deformation as might result from screening the powder will appreciably increase the intensity of the hexagonal lines.

In other words, it is rather difficult to obtain cubic cobalt entirely free from the hexagonal modification. If, however, one is so fortunate as to produce metal powder of the face-centered cubic form only, then after heating again in a hydrogen atmosphere at temperatures above 1200 degrees Cent. (2190 degrees Fahr.) for long periods (up to 100 hours) and cooling to room temperature at a moderate rate, the powder yields a pattern showing distinctly some hexagonal reflections.

These seem to become more intense as the time at the elevated temperature is extended. However, I have no data of a quantitative nature to bear this out.

It would seem that the question can be answered only by obtaining a diffraction pattern at the high temperature from cobalt which has been held previously at this temperature for a long period since the change in lattice form appears to go on very slowly.

Oral Discussion

R. H. THIELEMANN:[1] During your investigations did you notice any change in physical properties of the alloys with the occurrence of the order to disorder transformation? That is, do the iron-cobalt alloys become more brittle when they are in the ordered state? When you detected the time for the alpha to gamma transformation to occur for the iron-cobalt alloy containing 50.3 per cent cobalt, I noticed that a fairly long period was required on heating for complete transformation. I think Bain in his "Functions of the Alloying Elements in Steel" published by the A. S. M. showed that on cooling, an iron-cobalt alloy containing 10 per cent of cobalt and about 0.90 per cent of carbon transformed very rapidly. The knee of the S-curve was shifted so far to the left by the cobalt addition that only pearlite and ferrite were formed from austenite. The alloy was thus rendered almost nonhardenable by heat treatment, and this would indicate the occurrence of a small volume change during the transformation reaction. Since you showed a fairly long time for the transformation to occur on heating, I would like to know if you also found a large volumetric dilation accompanying the transformation reaction?

F. C. KELLEY:[2] I should like to ask Mr. Ellis what the degree of purity of his cobalt was. It seems in an investigation of this kind that the percentage of nickel in the cobalt should be very low and it is very difficult to make this nickel-cobalt separation.

G. F. GEIGER:[3] I would like to know if the authors observed any change in volume during the order-disorder change; if so, in which direction and about what magnitude?

S. R. HOOD:[4] In our work on thermal expansion we have found a general relationship between lattice constants and expansion rate. Since this investigation shows a large variation in lattice constants it would be interesting to know if the authors have information on the expansion rates of these alloys.

Authors' Reply

Dr. Sykes' discussion emphasized the limitations and scope of the present paper, that is to cobalt contents to 90 per cent. The authors did not include a study of the transformations in cobalt because such a study is an extensive investigation in itself. The work which has been done in connection with the high temperature transformation in cobalt, including the excellent work of Dr. Sykes, indicates the magnitude of the problem. The authors agree with Dr. Sykes that the present disagreement in the existence of the high temperature

modification of cobalt can best be confirmed or disproved by X-ray diffraction measurements at elevated temperature. Such measurements involve experimental difficulties because of the high temperatures, 800 to 1200 degrees Cent. but appear to have been successfully made by Marick (18).

The existence of a hexagonal phase in cobalt stable at high temperature is pertinent in connection with the gamma solid solubility limit as established by the authors. If the constitutional diagram should be such that the addition of iron rapidly depresses the disputed phase change there is the possibility of stable hexagonal phase at the equilibrating temperatures in certain of the alloys used in determining the lattice constant-composition curve for gamma alloys. This would lead to an error in the location of the phase boundary. Actually no reflections originating from a hexagonal structure were observed in the alloys containing 90 and 95 per cent cobalt, and moreover the lattice constant values were not sensitive to a change in the heat treating temperatures. Had the alloys in question been two-phase (gamma + hexagonal) it is almost certain that there would have been a change in the lattice constant of the gamma phase when quenched from different temperatures.

In reply to Mr. Thielemann's question, the 50 per cent alloy when slowly cooled, that is possessing a large degree of order, is appreciably more brittle than the same alloy quenched from above the critical temperature of order. Likewise, the electrical resistivity is greater for the quenched than for the slowly cooled alloy. Differences would be expected in most of the physical properties because of the difference in atom configuration in the two states.

Mr. Thielemann's discussion of the austenite decomposition in 10 per cent cobalt steel is interesting. We did not measure the volume changes accompanying the alpha-gamma transformations in this series of alloys. Masumoto (8) made such measurements and has reported a contraction accompanying the alpha→gamma transformation in the 10 per cent cobalt binary alloy of a magnitude somewhat larger than that of iron reported in the same paper. I think there would be some question as to whether results obtained for binary iron-cobalt alloys would be indicative of behavior in a cobalt steel where carbon could certainly alter the constitutional relationships to a marked extent.

Mr. Kelley has brought up the question of the nickel impurity in cobalt used in this investigation. Spectroscopic analysis showed no substantial amount of nickel. A wet analysis showed 0.04 per cent nickel.

Messrs. Geiger and Hood have both inquired with regard to the dilation in iron-cobalt alloys exhibiting the order-disorder transition. We have not made dilatometer analyses of these alloys and hence are unable to give a direct answer to the inquiries. It may be said safely, I believe, that anomalous expansion behavior would be expected in the temperature region below the critical temperature of order. Generally also there is a volume change at the critical temperature in the order-disorder transition. Lattice constant, measurements at room temperature on alloys quenched from the disordered and ordered regions do not provide data adequate to determine the magnitude and direction of the volume change at the transition. Such information could only be obtained by measuring length or volume change with temperature through the ordering region with rather slow rates of heating and cooling to ensure approximate attainment of equilibrium at the temperatures in question.

FATIGUE AND DAMPING STUDIES OF AIRCRAFT SHEET MATERIALS: DURALUMIN ALLOY 24ST, ALCLAD 24ST AND SEVERAL 18-8 TYPE STAINLESS STEELS

By R. M. Brick and Arthur Phillips

Abstract

The strong wrought aluminum alloys, 24ST and Alclad 24ST, and nine 18-8 type commercial grades of stainless steels of low and moderate carbon contents, and in the annealed, cold-rolled, stabilized and aged conditions were fatigue tested under constant deflection conditions. The S-N data so obtained indicate the probable endurance limits but are incomplete because of lack of time for aluminum alloys and the scatter of results for many of the stainless steels. The ratio of endurance limit to tensile strength of the stainless steels was found to be relatively constant for one type of surface condition but to change a maximum of from 30 to 70 per cent with different commercial finishes. Errors considered were mechanical or inherent in the test method; i.e., increases of stress at constant deflection near the endurance limit accompanied by plastic behavior resulted from work hardening or decreases in apparent stress from crack formation and propagation, particularly in Alclad. The effects of the two opposing factors, work hardening and cracking, were studied by means of damping and mechanical hysteresis (load-bending) tests on actual fatigue specimens at various intervals during stressing above and near the endurance limit. Very considerable changes took place in Alclad specimens but changes in the stainless steels were relatively slight although considerable differences were found among the various types studied.

QUANTITATIVE data on the fatigue and damping properties of these materials are available in the literature but differences in test methods and gaps in the published information make correlation difficult and obscure the significance of the results. For example, there are no published values on the fatigue strength of Alclad 24ST

although the NACA has published the results of fatigue tests on Alclad sheet duralumin 17ST (1)[1] and it might be assumed that the former, somewhat stronger, alloy would have a slightly higher endurance limit. Numerous data on the fatigue properties of 18-8 stainless steel have been published but only one report (2) was based on reversed bending of sheet metal. The present work embraces this type of fatigue testing of 24ST, Alclad 24ST and 18-8 Cr-Ni steel (3) in the annealed condition, cold-rolled to various degrees, and with differences in carbon content and the addition of carbide stabilizers.

MATERIAL

The strong wrought aluminum alloy 24S is one of the so-called super-duralumins with the following nominal chemical composition: copper 4.4 per cent, magnesium 1.5 per cent, manganese 0.6 per cent, balance aluminum. Iron and silicon are always present and the silicon is considered to be an essential element in the amounts usually present. After a solution heat treatment, the alloy age hardens spontaneously at room temperature. The heat treated and aged condition is denoted by the temper designation "T". The alloy is available in the form of sheet with a surface coating of pure aluminum and is then known by the trade name, Alclad.

The alloys Alcoa 24ST and Alclad 24ST, in the form of sheets of approximate dimensions 14 by 20 by 0.032 inches, were procured from a prominent aircraft manufacturer. Careful measurements of the gage of all sheets disclosed that the actual thicknesses varied from 0.0305 to 0.0332 inch. Specimens for tensile and fatigue tests were cut from the sheet with the direction of rolling both transverse and parallel to the axis of the test specimens. The rolling direction indicated by the surface markings on the sheets was verified by the alignment of insoluble constituents of the alloy as observed by microscopic studies of polished sections. By microscopic observations, on sections transverse to the rolling plane, the thickness of the originally pure aluminum layer on the Alclad sheet was found to vary from 0.0015 to 0.0016 inch. The examination also showed that the material had been given a normal solution treatment. No melted eutectic was found as in "burnt" aluminum alloys and only a small amount of undissolved $CuAl_2$.

The stainless steels were obtained in sheets of the same gage, in

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[1]The figures appearing in parentheses refer to the bibliography appended to this paper.

most cases, as the aluminum alloys. The chemical analyses and physical conditions of the materials are given in Table I and their mechanical properties together with those of the aluminum alloys in Table II.

Table I
Chemical Analyses and Physical Conditions of the 18-8 Cr-Ni Stainless Steels

Specimen No.	Type	C	Ni	Cr	Mn	P	S	Si	Surface Finish	Physical Condition
1	KA2S;304	0.08	8/10	18/20	(nominal analysis)				No. 4	Annealed
2	KA2S;304	0.08	8/10	18/20	(nominal analysis)				No. 4	Cold-rolled about 10 Per Cent
3	18-8;304	0.07	8.69	18.67	0.40	0.011	0.008	0.40	No. 2-B	Annealed
4	18-8;302	0.11	8.88	18.03	0.57	0.022	0.007	0.41	No. 2-B	Annealed
5	18-8;304	0.06	8.63	17.90	0.43	0.012	0.008	0.36	No. 2-B	Cold-rolled about 50 Per Cent
6	18-8;302	0.115	8/10	18/20	—	—	—	—	Rolled	Cold-rolled about 50 Per Cent
7	18-8 Cb;347	0.06	9.18	17.36	0.47	0.015	0.010	0.40	(.86 Cb) No. 2-B	Note *a*
8	18-8;302	0.13	7.49	18.00	1.48	0.013	0.017	0.27	No. 2-B	Cold-rolled about 50 Per Cent
9	18-8 special	0.065	9.13	17.49	(No complete analysis available although other elements must be present to account for properties)				Rolled	Cold-rolled and aged

Note *a*—Sheet rolled to 130,000 pounds per square inch tensile and aged 1 hour at 850 degrees Fahr., raising strength to 185,800 pounds per square inch.

Specimens Nos. 1 and 2 were supplied by the Crucible Steel Company, through the courtesy of Mr. F. J. Dawless.

Specimens Nos. 3 to 8 were from the Carnegie-Illinois Steel Corporation. Analyses were supplied by Mr. H. V. Thaden of that Company.

Specimen No. 9; an experimental heat exhibiting unusual age hardening capacity. Alloy cold-rolled and aged one hour at 850 degrees Fahr.

Usual commercial surface finishes: No. 2-B; hot rolled, annealed, pickled, cold rolled, annealed, pickled, given bright cold-rolled surface. No. 4; same as No. 2-B followed by a rough grind and then a polishing operation. Finish listed as "Rolled" same as No. 2-B except that final rolls were not highly polished.

(Specimens Nos. 1-4 showed a slight response; Nos. 5, 6 and 8 a strong response; and Nos. 7 and 9 a very strong response to a magnet)

The data reported are the averages of three tests for each material. The 24ST specimens were found to be sensitive to surface markings and often broke at the scratch marking the gage length, thereby vitiating the elongation measurement but seldom affecting the breaking load; the Alclad 24ST was less sensitive to surface scratches and the stainless steels showed no effect of this type.

It is interesting to note that Alclad 24ST showed tensile strength properties only 3 to 5 per cent below that of the regular 24ST in spite of the fact that 10 per cent of its cross section appeared to consist of pure aluminum. It seems probable that the explanation is to be found in a partial diffusion of copper and magnesium from the

alloy core into the adjacent layer of aluminum. That this occurred during rolling and heat treatment of the material is shown by the manner in which the cross section responded to the metallographic polishing operation, (Fig. 4).

Table II
Tensile Properties of Materials

Specimen	Y.P (a) Pounds Per Square Inch	T.S. Pounds Per Square Inch	Elong. Per Cent In 2 Inches	Rockwell Hardness	Yield Point Pounds Per Square Inch	Tensile Strength Pounds Per Square Inch	Elong. Per Cent in 2 Inches	Rockwell Hardness
		Test Values				Nominal Values*		
24ST-L(b)	44,000	65,700	20.	F97	44,000	68,000	19	...
24ST-T(c)	41,000	65,000	18.	F97				
Al.24ST-L	43,000	64,300	20.	...	41,000	62,000	18	...
Al.24ST-T	42,000	62,500	17.5	...				
(1) 18-8;KA2S-L	40,000	83,500	57.	B80	....	....	..	...
(1) 18-8;KA2S-T	40,000	84,000	60.	B80	....	....	..	...
(2) 18-8;KA2S-L	53,000	95,000	47.	B92	....	....	..	...
(2) 18-8;KA2S-T	59,000	96,000	45.	B92	....	....	..	...
(3) 18-8;304	26,000	75,000	64.	B77	45,000	88,000	56	B81
(4) 18-8;302	28,000	89,000	65.	B76	49,000	98,000	57	B82
(5) 18-8;304	95,000	158,000	16.	C31	152,000	168,000	13	C35
(6) 18-8;302	105,000	180,000	10.	C40	....	....	..	...
(7) 18-8Cb;347	125,000	194,000	2.5	C40	192,000	198,000	3	C41
(8) 18-8;302-L	135,000	195,000	8.	C41	200,000	223,000	8	C44
(8) 18-8;302-T	135,000	213,000	8.7	C41				
(9) 18-8 Special	175,000	243,000	1.5	C45	....	....	..	...

*Nominal values for aluminum alloys taken from A.S.M. Metals Handbook.
For stainless steels, they are values quoted by the supplier of the material.
(a) Yield Point; By dividers for duralumin and stainless (1) and (2); stress producing permanent set of 0.1 per cent for other stainless steels, (Huggenberger Tensometer).
(b) L—longitudinal specimen, axis parallel to rolling direction.
(c) T—transverse specimen, axis transverse to rolling direction.

FATIGUE TESTS

Testing Method—Fatigue test specimens of the shape shown in Fig. 1 were cut with a special milling cutter. The 24ST, Alclad 24ST and stainless steels, Nos. 1, 2 and 8 were cut with axes of the specimens both transverse and longitudinal, i.e., across and parallel to the rolling direction of the sheet while other stainless steels were tested only as longitudinal specimens. The milled edges were polished with No. 1 emery paper, then with No. 00 emery paper.

As indicated in the sketch, the specimen has a section with hypothetical constant bending moment. To determine the stress-deflection relationships for the different specimens, they were set up in a holder similar to that of the actual testing machine and increments of load applied in a horizontal direction. The deflection corresponding to a given load, applied at the end of the hypothetical beam of constant moment, was measured with a microcomparator to the nearest 0.001

inch. The maximum fibre stress for a given deflection may be calculated from the equivalent load by use of the familiar beam formula;

$$S = \frac{6\,Pl}{bd^2} \text{ where}$$

S = stress in psi,
P = load in pounds,
l = beam length, inches,
b = beam width, inches,
d = beam thickness, inches.
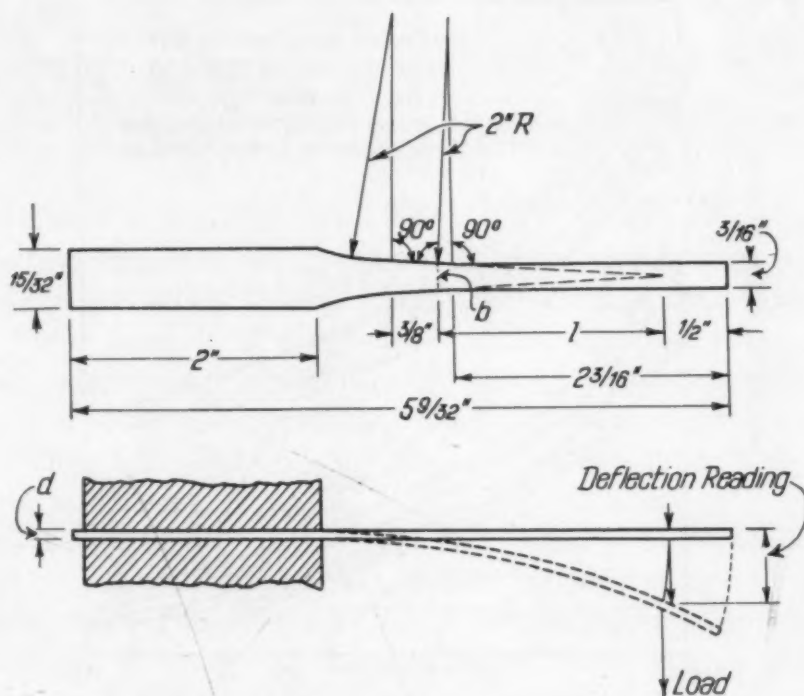
Fig. 1—Fatigue Test Specimen and Method of Determining the Load-deflection Relationship.

The fatigue testing machine was a constant deflection type identical to that described by Townsend and Greenall (4). It accommodated 40 specimens, four different alloy groups of five specimens at two different chosen deflections. The machine was operated at a speed of 900 revolutions per minute. Automatic telephone relays to individual counters recorded the number of cycles required for fracture of each specimen.

The tests were started at high deflections or stresses and then proceeded into lower stress ranges approaching the endurance limit of the materials. With only two deflections simultaneously available on the machine, it becomes a lengthy procedure to determine the terminus of the S-N curves for the several different materials, each requiring a different deflection at the final fatigue life and perhaps

several months operation at that deflection to establish the results with a reasonable degree of certainty.

Results; Aluminum Alloys—The results of the tests on aluminum alloys 24ST and Alclad 24ST are presented graphically in Fig. 2. Since the number of specimens available from the original supply of sheet was inadequate, a second group of sheets were ob-
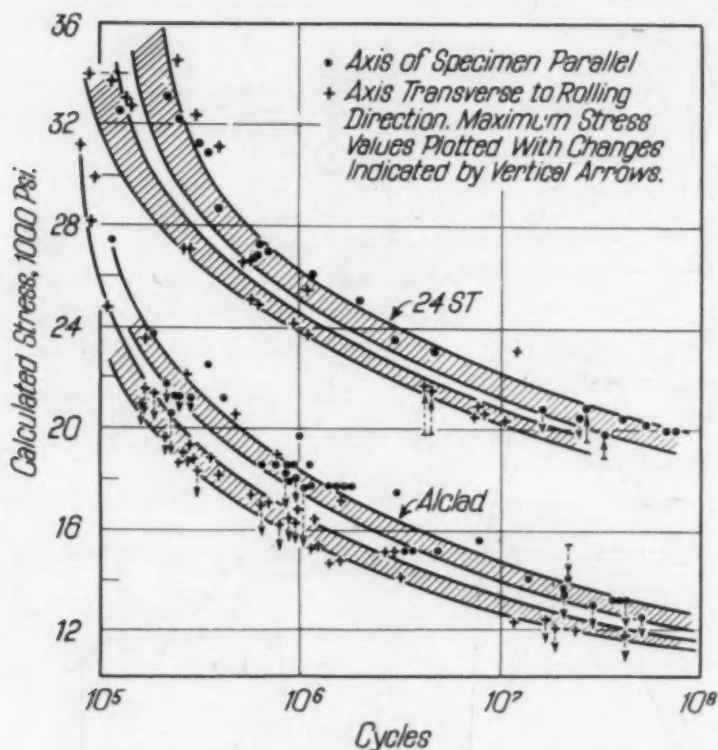
Fig. 2—Fatigue Data for Duralumin 24ST and Alclad 24ST.

tained at a later date. In the case of the original sheets, several specimens gave fairly concordant load-deflection tests and an average curve for each type of specimen was determined for use in calculating the stresses involved for each deflection setting of the testing machine. On testing specimens from the second group of sheets, a considerable scatter in results forced a more careful examination of individual specimens. A large variation in load-deflection characteristics was found, usually associated with a variation in gage of the specimens. Thenceforth, a load-deflection curve was determined for each individual specimen tested, and in Fig. 2, individual datum points are plotted for most of the tests. Time did not permit the completion of tests to the ordinarily accepted limit of 500,000,000

cycles of stress for aluminum alloys. From the S-N curves shown, the following results are suggested:

Material	Fatigue Strength 50,000,000 Cycles Pounds Per Square Inch	Probable Endurance Limit Pounds Per Square Inch	Limit by Other Data Pounds Per Square Inch
24ST—L[1]	20,500[3]	18,000[4]	18,000[5]
24ST—T[2]	18,500[4]	16,500[4]	
Alclad—L	13,000[3]	11,000[4]	10,000[6]
Alclad—T	11,500[3]	10,000[4]	

(1) L—rolling direction parallel to specimen axis.
(2) T—rolling direction transverse to specimen axis.
(3) Error from stress measurement and extrapolation; ± 1000 pounds per square inch.
(4) Error; ± 2000 pounds per square inch.
(5) Data from National Metals Handbook.
(6) Data on Alclad 17ST from National Bureau of Standards' tests; 17th Annual Report, National Advisory Committee for Aeronautics. Error ± 10 per cent.

Discussion of Results; Aluminum Alloys—Two types of errors are encountered in the determination of the stress created in individual specimens when deflected a given amount in the testing machine. First, there is always a considerable inherent error where tests are carried out under constant deflection conditions. The stresses of tension and compression set up in the outer fibers of the specimen during the first cycle are not constant but change during the life of the specimen.

The change found in the 24ST in most instances, particularly in the vicinity of the endurance limit, was an increase in stress for constant deflection. The process of cyclic stressing seemed to result in an absorption of energy by the specimen. Some energy was dissipated in the form of heat but a considerable amount apparently was stored in the metal in a way perhaps analogous to the increase in energy found upon cold deformation. The result was an increase in stiffness or in stress at a constant deflection. Fig. 2 shows that the maximum increase in stress found in these specimens appeared to be in the neighborhood of 2000 pounds per square inch. It seemed most logical to employ the maximum stress found for each specimen in plotting the graph.

X-ray diffraction patterns of the surface layers were made to determine whether the work or energy stored in the 24ST would be manifested. Back reflection photograms of unstressed specimens of 24ST gave spotty diffraction circles composed of reflections from individual crystals. After stressing above the fatigue limit to fracture of the specimen, the diffraction spots changed to form a somewhat diffuse continuous ring, in a way analogous to a cold-worked material. However, a similar reflection pattern was obtained on

unbroken specimens after 88,000,000 cycles at stresses near or below the endurance limit. This result verifies previous work by Barrett (5) which has demonstrated that X-ray studies may indicate crystal fragmentation during fatigue but are incapable of showing incipient failure.

In the Alclad specimens, an increase in stress was noticed in a few specimens but as is evident from Fig. 2, an apparent decrease in stress was usually found. If specimens had been removed from the testing machine more frequently for load-deflection determinations of stress, a clearer definition of the changes might have been possible. However, the photomicrographs of Figs. 3 and 4 offer a positive explanation of the decrease in stiffness or calculated stress for the Alclad specimens. These micrographs were taken in the section of maximum stress of a longitudinal Alclad specimen unbroken after 900,000 cycles at a deflection of 0.46 inches (original stress, 18,200 pounds per square inch; stress after the 900,000 cycles, 17,400 pounds per square inch). Evidently the aluminum layer became interlined with transverse cracks during cyclic stressing. The cracks reduced the effective thickness of the beam and decreased its stiffness and thus the calculated stresses.

X-ray back-reflection diffraction studies of the Alclad during fatigue gave results somewhat different than the 24ST. Reflection circles from the pure aluminum surface were weak while those from the underlying alloy core were in all cases quite strong, continuous in character and apparently unchanged by cyclic stressing. A method ensuring reflection chiefly from the surface might have shown changes in the aluminum layer but it was not possible to study this in detail.

The incidence and propagation of surface cracks in the Alclad sheet was a result of the low fatigue strength of the relatively pure aluminum surface layer. The same characteristic, low surface strength of the Alclad, may cause an additional inherent error when the beam formula is applied in calculating the maximum surface stress in a deflected specimen. Dix, in discussing a paper by Rawdon (6) on corrosion fatigue in duralumin and Alclad, stated that application of the simple beam formula to a bi-metal material like Alclad might lead to a 25 per cent error in the calculated stress.

The second inherent error arises from stresses originally present in the age hardened sheet which caused some of the milled specimens to assume a slight bow or curvature. Since such specimens were not at zero stress at the neutral point, the maximum stress caused by
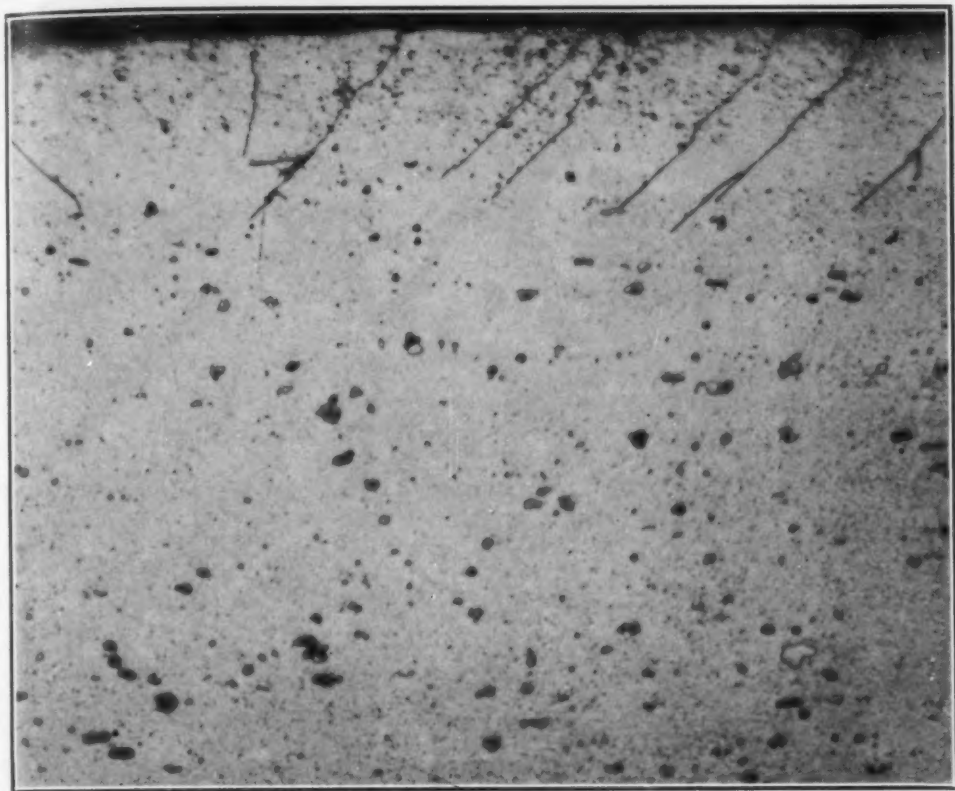
Fig. 3—Fatigue Cracks in an Unbroken Alclad 24ST Specimen Subjected to 900,000 Cycles at 18,200 Pounds Per Square Inch; Showing Their Frequency and Characteristic Angle to the Surface. Core Shows Normal Structure of 24ST; Alclad Layer Shows a Zone, Polished More Cleanly, Where Cu and Mg Have Diffused From the Core. Etch, 0.5% HF. × 500.

equal deflection in either direction from the neutral point was necessarily greater on one side than the other.

Another type of error affecting the results of this work, such as variations in gage with considerable resultant differences in load deflection data, were subject to some control or analysis. Stress values used in Fig. 2 were based on individual load-deflection tests and careful dimensional measurements of each specimen. In spite of this, the calculated stresses only approximate the true values for the following reasons.

1. The depth of the beam of constant bending moment as the specimen was mounted in the testing machine could not be measured with complete accuracy.

2. The actual length of the beam in the testing machine varied.

3. As a result of errors 1 and 2, the l/b ratio of the beam stress formula, which should theoretically be 9.269, actually varied from 8.90 to 9.40.

Fig. 4—Same Specimen as Fig. 3 Showing Progress of Fatigue Crack from Aluminum Layer Into Alloy Core. Etch, 0.5 Per Cent HF. × 1300.

Not all but many of the data include specimens on which there was at least an approximate knowledge of the above errors. A plot of fatigue life vs. machine deflection, for example, gave a curve with widely scattered points. The calculated stress values plotted against the fatigue life of specimens still showed some scattering of points but to a much lesser degree. The errors given in the table of results (previous section) are estimates of the limits.

A final pertinent factor, corrosion, has not been considered in this work but would inevitably influence the use of fatigue data in engineering design. It has been found (Rawdon, loc. cit.) that the endurance limit of the duralumin type alloys is greatly lowered by simultaneous stress and corrosion. The intercrystalline attack of salt solutions or atmospheres may decrease the fatigue strength of 17ST by 50 per cent. Yet Alclad, in spite of the cracks early developed in the aluminum layer, is electrolytically protected by that layer and its endurance limit is believed to remain virtually unchanged.

It appears from the results of the present work that Alclad 24ST

would be more adversely affected by overstressing than the 24ST due to the incidence of cracks and their action in localizing stress conditions. It may be suggested that endurance tests conducted on a constant-load basis, instead of the constant-deflection system here used, would probably show a still lower endurance limit for Alclad. The effect of a weaker surface layer greatly decreasing the fatigue strength of a strong alloy is not confined to Alclad but is probably generally true. Boegehold (7) has cited the strongly adverse effect of a decarburized skin on the fatigue strength of heat treated steels and vice versa, the increase in endurance limit of carburized steels.
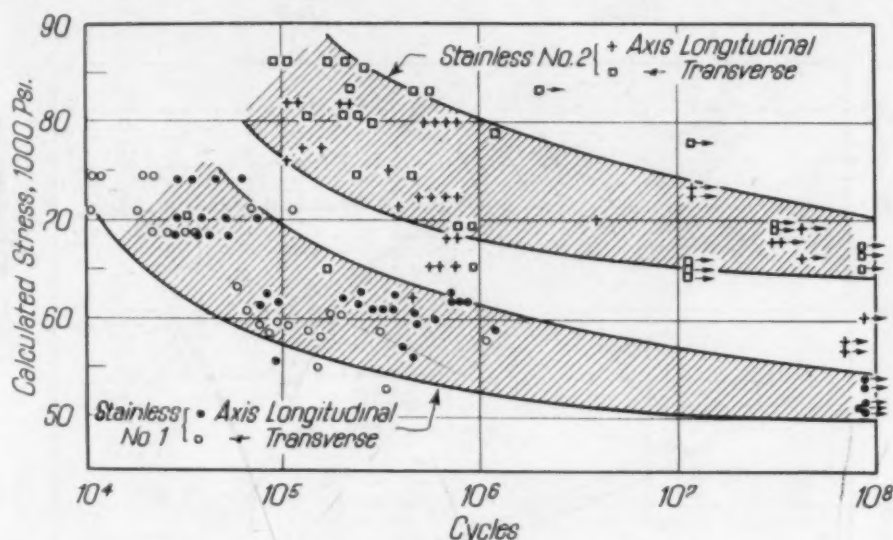
Fig. 5—Fatigue Data for Stainless Steel No. 1 (Type 304, Annealed) and Stainless Steel No. 2 (Type 304, Slightly Cold-Rolled).

Results; Stainless Steels—The fatigue data relating to stainless steels Nos. 1 and 2 are reproduced in Fig. 5. The results show considerably greater scatter than did the aluminum alloys. Despite the width of the zones representing the S-N curve, a few points are outside these generous limits. Examination of the figure discloses that in almost all cases, if an individual specimen did not break in less than one million cycles, it withstood the applied stress indefinitely, (lasted more than 10 million cycles, in some cases up to 90 million).

The inconsistency of the results led to an intensive examination of the sources of error discussed previously. Since the steel blocks holding the reciprocating end of the specimen showed marked signs of wear, new carburized and hardened blocks were inserted. A tabulation of the specimen sites did not appear to account for the variations noted. Load-bending data obtained on specimens at vari-

ous stages during fatigue testing revealed that, as in the case of duralumin 24ST, the stress sometimes increased as much as 10 per cent during the run. This was particularly true of specimens stressed near the supposed endurance limit where for one specimen, the following stress increases were noted:

Stainless No. 2—Tested at 0.495 Inch Deflection

1000 Cycles	1000 Pounds Per Square Inch
Original condition	51.8
5	52.0
20	54.0
50	56.1
200	56.4
800	56.4
4760	56.9
21000	56.9
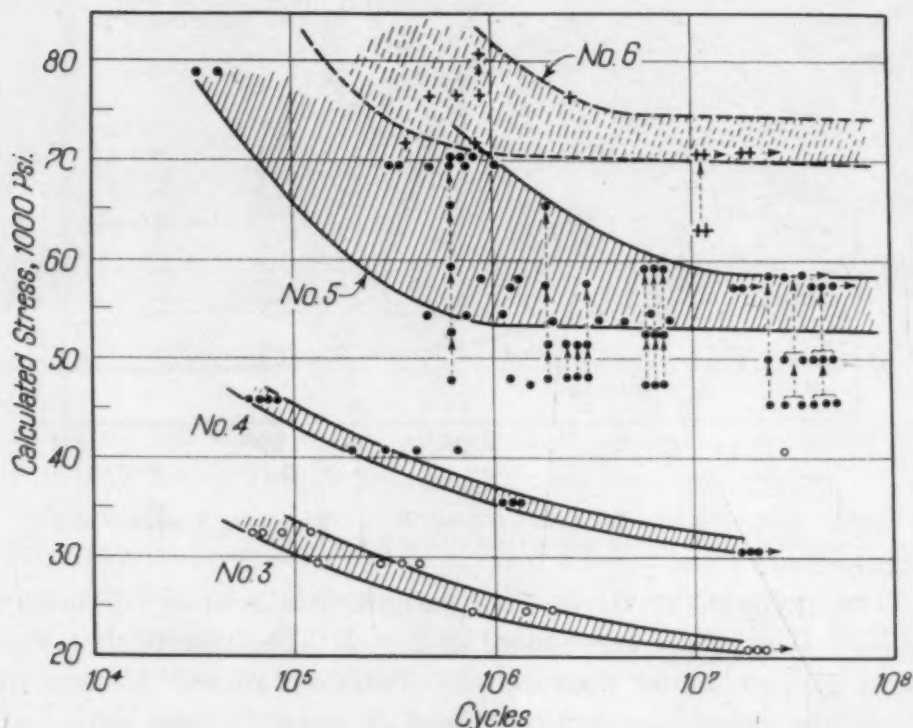
Fig. 6—Fatigue Data for Stainless Steels No. 3 (Type 304, Annealed), No. 4 (Type 302, Annealed), No. 5 (Type 304, Cold-Rolled) and No. 6 (Type 302, Cold-rolled). Vertical Arrows Signify 10,000,000 Cycles at the Lower Stress; the Stress Was Then Raised Successively Until the Specimen Broke at the Indicated Number of Cycles at the Highest Stress or Remained Unbroken (Horizontal Arrows).

In only a few instances were stress decreases noted and it seems logical to assume that these were a result of faulty measurement or of the presence of cracks which it will be recalled in the case of Alclad, resulted in an apparent stress decrease. The spread in the S-N data may be interpreted to signify that the early stages of stressing generally result in work hardening, an increase in stiffness

and consequently in the calculated stress. However, a slight surface defect may cause a crack to form in which case failure rather than strengthening results.

Most of the data on the other stainless steels are equally irregular, Figs. 6 and 7. Vertical lines in these figures indicate the
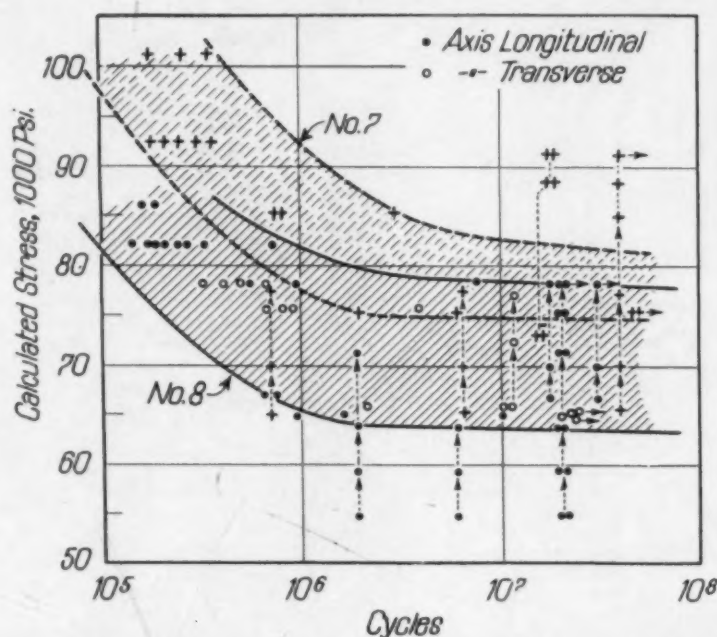
Fig. 7—Fatigue Data for Stainless Steels No. 7 (Type 347 Cb, Cold-Rolled and Aged) and No. 8 (Type 302, Severely Cold-rolled). Vertical and Horizontal Arrows Signify Same as in Fig. 6.

specimen was first stressed a minimum of 10,000,000 cycles at the lowest plotted value, the stress raised to the next value for 10,000,000 cycles and so on. The vertical line indicates the number of cycles to fracture at the highest stress only or the customary arrow to the right indicates failure did not occur. No data are shown for alloy No. 9 which was in too thin a gage to be adequately tested. A very much higher deflection was required to obtain a stress equivalent to other specimens. However, the endurance limit-tensile strength ratio seemed to be in the vicinity of 40 per cent. Changes in stress during the tests were found to occur in about the same degree as for Nos. 1 and 2. The endurance limits suggested by the curves are given in Table III with other published data for comparison. It should be emphasized here that apart from other errors, the quoted stress values are all calculated on the basis of purely elastic deformation in the cantilever beam specimens and that actually some plastic

deformation was always encountered with a resultant error in the calculated stress.

The data indicate that the ratio of endurance limit to tensile strength is only imperceptibly affected by the variables of carbon

Table III
Results of Fatigue Tests of Stainless Steels

	Alloy No.,	Type and Condition	Tensile Strength Pounds Per Square Inch	Indicated Endurance Limit Pounds Per Square Inch	Ratio Endurance Limit By Tensile Strength	Surface Finish
1	304	Annealed	84,000	50-55,000	59-65	No. 4
2	304	Cold rolled 10 per cent	96,000	65-70,000	68-73	No. 4
3	304	Annealed	75,000	22-24,000	29-32	No. 2-B
4	302	Annealed	89,000	32-34,000	36-38	No. 2-B
5	304	Cold rolled 50 per cent	158,000	53-58,000	33-37	No. 2-B
6	302	Cold rolled	180,000	70-74,000	39-41	Rolled
7	347Cb	Cold rolled and aged	194,000	75-81,000	39-42	No. 2-B
8	302	Cold rolled 50 per cent	195,000	63-78,000	32-40	No. 2-B
9	Special	Aged	243,000	....	.....	Rolled

Previously Published Data

Ref. No.*	Material & Test	T. S.	E. L.	Ratio
9	18/8 Rotating beam	82,800	36,000	43.5
9	18/8 Notched rotating beam	82,800	43,000	51.9
10	18/8 Rotating beam	88,000	33,000	37.5
10	18/8 Rotating beam	112,000	62,000	55.4
10	18/8 Rotating beam	127,000	62,000	48.8
9	18/8 Rotating beam	132,500	70,000	52.8
9	18/8 Rotating beam	132,500	49,000	37.0
10	18/8 Rotating beam	135,000	74,000	54.8
11	18/8 Haigh direct	148,000	56,000	37.8
11	18/8 Cantilever beam	148,000	53,000	35.8
2	18/8 1.5 Mo Reverse bend, sheet	212,000	85,000	40.0

*Reference numbers correspond to those of Bibliography.

content, degree of cold work and aging but that it is chiefly a function of the surface finish. All specimens with the commercial finish known as No. 2-B showed values for this ratio in the range of 30 to 40 per cent. Cold rolling seemed only to increase the ratio from the low to the high side of the range. Microscopic observations of all surfaces disclosed that in the case of the annealed stock, the grain boundaries had been deeply attacked during the final pickling operation and that the subsequent "bright roll" pass had not in reality given a bright finish nor affected the outlined boundaries, Fig. 8b. Severe cold rolling, following the final pickle, to the same listed No. 2-B finish, gave a bright, smooth surface but did not obliterate the etched grain boundaries, Fig. 8c, and also did not materially increase the endurance ratio. When the nominal No. 2-B finish was followed by a rough grinding operation and then a polishing operation to a commercial No. 4 finish, as in the case of the Crucible Steel Com-

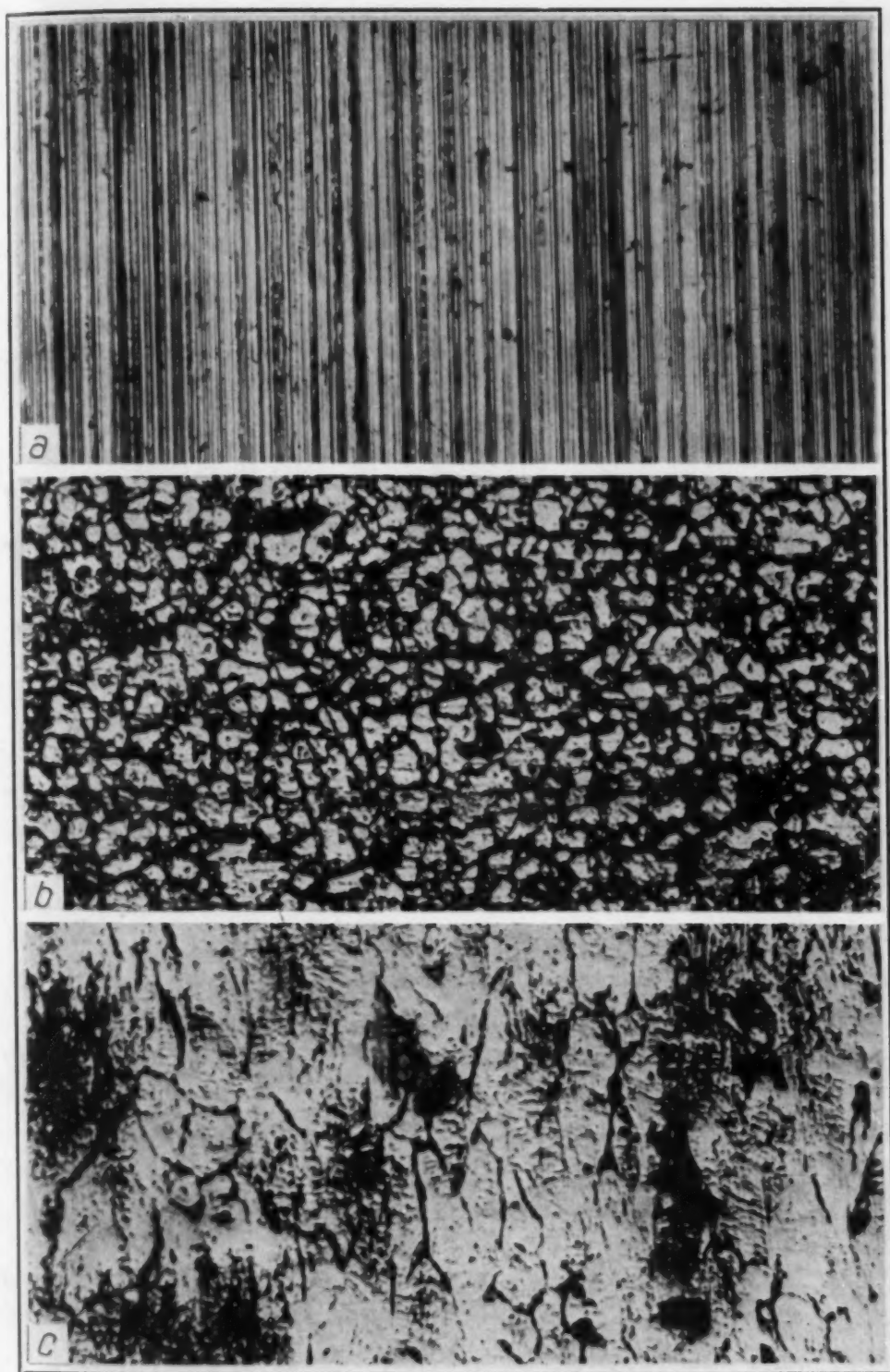Fig. 8—Micrographs of the Commercial Finishes on the Stainless Steels Specimens as Received and Tested; no Polish or Etch. × 200. (a) Finish No. 4, specimen No. 1 (304; Annealed), Endurance-tensile Ratio of 60-65 Per Cent; (b) Finish No. 2-B, Specimen No. 3 (304; Annealed), Endurance-tensile Ratio of 29-32 Per Cent; (c) Finish No. 2-B, Specimen No. 8 (302 Severely Cold Rolled), Ratio of 32-40 Per Cent.

pany specimens, the endurance ratio jumped from the 30 to 40 per cent to a 60 to 70 per cent range. This treatment not only removed all boundary etch effects, Fig. 8a, but subjected the most highly stressed part of the fatigue specimen to an efficient localized work hardening operation.

None of the fatigue data can be taken to represent the true endurance limits of the stainless steels but show rather those of the surfaces involved. It was hoped that the finishes listed in Table I as "Rolled" would be more representative of the material but even these specimens showed strong evidences of grain boundaries from previous pickling operations. The theoretical endurance limits are, however, less important commercially than those actually obtained on a given material. Where vibration of the part in service results in maximum stresses at the surface, a grinding operation would seem to be required to obliterate all etched boundaries which might initiate cracks and lower the fatigue strength actually achieved.

Damping Tests

The ability of metals to dissipate vibrational energy in the form of heat is of considerable importance in many engineering services. Under cyclic stresses either above or slightly below the fatigue limit, most metals exhibit a combination of elastic and plastic deformation and in general the ratio of plastic to elastic strain is an indication of the "damping capacity" of the material.

A brief review of the literature on damping suggests the probable importance of such tests made in conjunction with fatigue studies. Brophy (19), (20) has shown that the damping capacity of carbon steels seem to be related to the damage resulting from overstress and notches. Förster and his co-workers (13), (14), (15) have shown that mechanical cracks and strain-hardening each greatly affect the damping capacity of a metal as determined by resonance measurements. Furthermore, they found that damping of Al-Mg alloys was little affected by general corrosion but greatly increased by intercrystalline corrosion. These data, and others listed in the bibliography (16), (17), suggest that fatigue cracks, of the type shown in Figs. 3 and 4, should markedly affect the damping capacity of the specimen and that stress changes accompanying work hardening during fatigue stressing should also result in a change in damping.

A specimen of 24ST carried through a complete cycle of stress may exhibit a load-bending curve of the type shown in Fig. 9. This

is not a true stress-strain diagram but plotted load and displacement values of the cantilever beam are proportional to the stress and strain in the outer fibers of the fatigue specimen. Increasing and decreasing loads in both directions followed identical lines; no plastic strain was
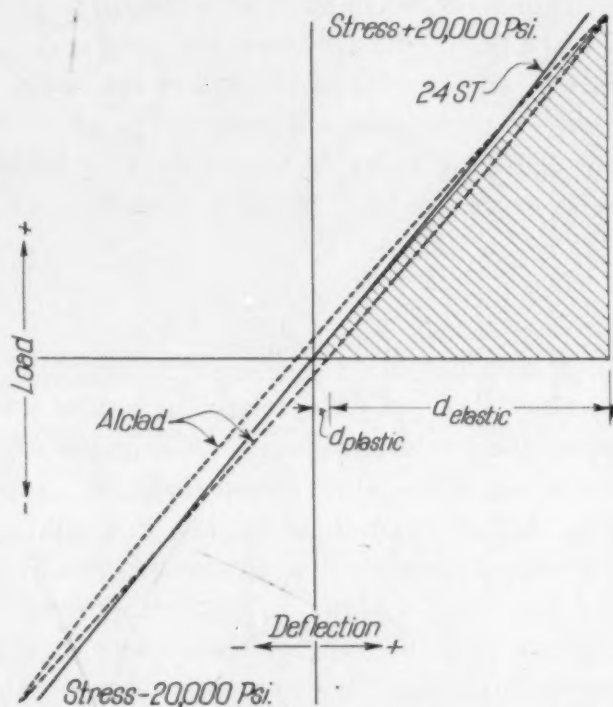
Fig. 9—A Complete Load-bending Cycle for Specimens of Duralumin 24ST and Alclad 24ST Showing Mechanical Hysteresis for the Latter.

involved in the deformation and for that particular maximum stress of about 20,000 pounds per square inch, the damping capacity must be exceedingly low. As a matter of fact, no measurable plastic deformation was found for 24ST with normal rates of loading until the maximum stress exceeded 30,000 pounds per square inch.

On the same graph, the dotted line shows the load-deformation curve obtained on an Alclad fatigue specimen taken through the same cycle of loading. It will be observed that a definite permanent set in the specimen indicated some plastic deformation occurred during loading in each direction. The total cycle of load versus bending resulted in the formation of a mechanical hysteresis loop. Such a hysteresis loop was found for Alclad specimens even at loads equivalent to stresses of 7000 pounds per square inch, well below the endurance limit of the material. A similar hysteresis loop has been obtained after prolonged cyclic stressing of steels and indeed there is

some evidence, according to Föppl (17), that the ultimate damping, i.e., the steady value attained after some millions of cycles, may be considerably greater than the initial damping.

The damping capacity can be evaluated as a non-dimensional figure in terms of the mechanical hysteresis loop of Fig. 9 by:

(1) the ratio λ of the plastic strain d_{pl}, to the elastic strain d_{el}; or (2) the ratio Ψ of the area of the hysteresis loop to the work of deformation stored in the extreme position, (diagonally hatched). Accordingly, the damping capacity for 24ST and Alclad at a stress of 20,000 pounds per square inch, has been calculated from Fig. 9 as:

	λ	Ψ
24ST	0	0
Alclad	0.10	0.14

The above values show only the relatively greater damping capacity of Alclad since the hysteresis loop used to calculate λ and Ψ has not been plotted as a definite stress-strain diagram. (In spite of the purely qualitative definition of these values, it should be noted that they are in fairly good agreement with the quantitative values later determined; cf. Fig. 14.) Many other methods for determining damping capacity are available but for these sheet fatigue specimens, the only suitable means are those employing a freely vibrating cantilever beam. In most cases, although the shear stress is most important in damping as in fatigue, the damping found by bend-tests usually has a fairly constant relationship to the damping found by torsional oscillations or even by resonance methods.

Test Method—A heavy steel plate was firmly bolted to a concrete foundation pier, establishing a stable, vibrationless base. The specimens, actual fatigue specimens (Fig. 1) were securely held in a position similar to that employed during fatigue testing; i.e., the plane of vibration was horizontal. A steel block weighing approximately 26 grams was attached to the end of the specimen and served as an inertia weight. A small concave mirror attached to the block focussed a beam of light reflected from a stationary, single filament galvanometer bulb onto an 8 x 10 inch sensitized paper rotated on a 6-inch diameter cylinder in front of the vibrating beam. A magnetized block to one side of the free end of the specimen served to deflect it any chosen amount, corresponding to a definite initial stress. Cutting the current to the magnet released the specimen to vibrate freely while the decaying vibrations, magnified approximately

Several of the damping test films are reproduced here for illustrative purposes, Figs. 10, 11 and 12. The decrement was directly measured at various stages, and the ratio of decrement to the amplitude, or the derived damping capacity $\dfrac{\Delta W}{W} = \dfrac{2 \Delta A}{A}$ (ref. 12) was plotted against the calculated stress corresponding to the deflection at that cycle. The changes in damping of an Alclad specimen
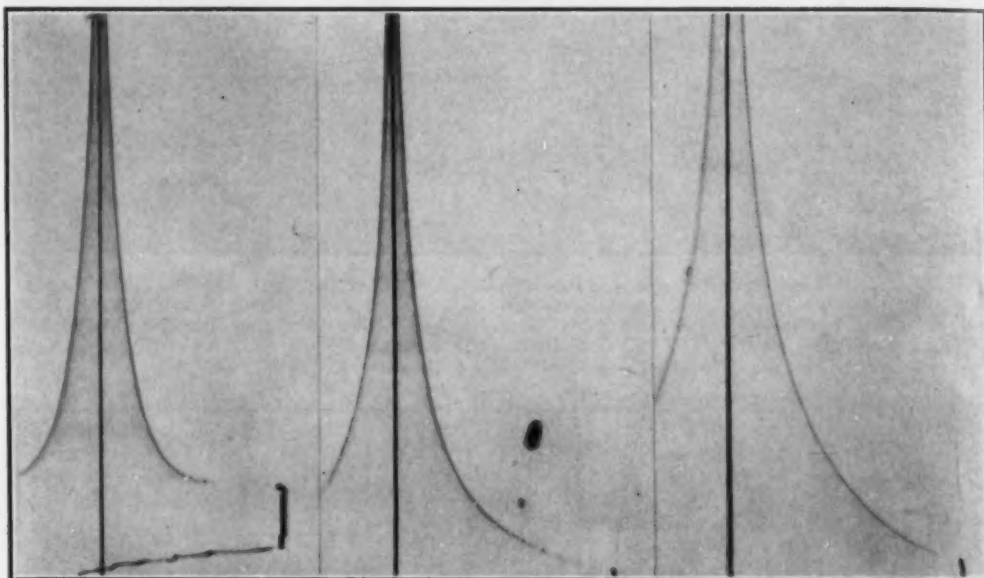
Fig. 12—Damping Curves of Unstressed Stainless Steels: (a) No. 4 (Type 302, Annealed)—Shows Permanent Set from Damping Test; (b) No. 5 (Type 304, Cold-rolled); (c) No. 8L (Type 302, Severely Cold-rolled).

during fatigue stressing well above the endurance limit are reproduced in Fig. 13, (cf. Fig. 9b and 9c). Load-deflection tests were made at each time the fatigue stressing was interrupted for a damping curve to be taken. The table in Fig. 13 shows how the calculated stress was changing at the same time as marked differences appeared in the damping curves. In the first 5000 cycles, a slight rise in the fatigue stress was accompanied by a decrease in damping, particularly at higher stresses. Next, a sharp decrease in stress occurred while the damping capacity continued to decrease at higher stresses but started to rise at low stresses. It is believed that this rise indicates that cracks were forming and propagating in the aluminum surface layer. Finally in the latter stages, shortly before the specimen broke (at 490,000 cycles), the damping capacity in the low stress range, 2000 to 10,000 pounds per square inch, considerably increased as the fatigue cracks penetrated the alloy core.
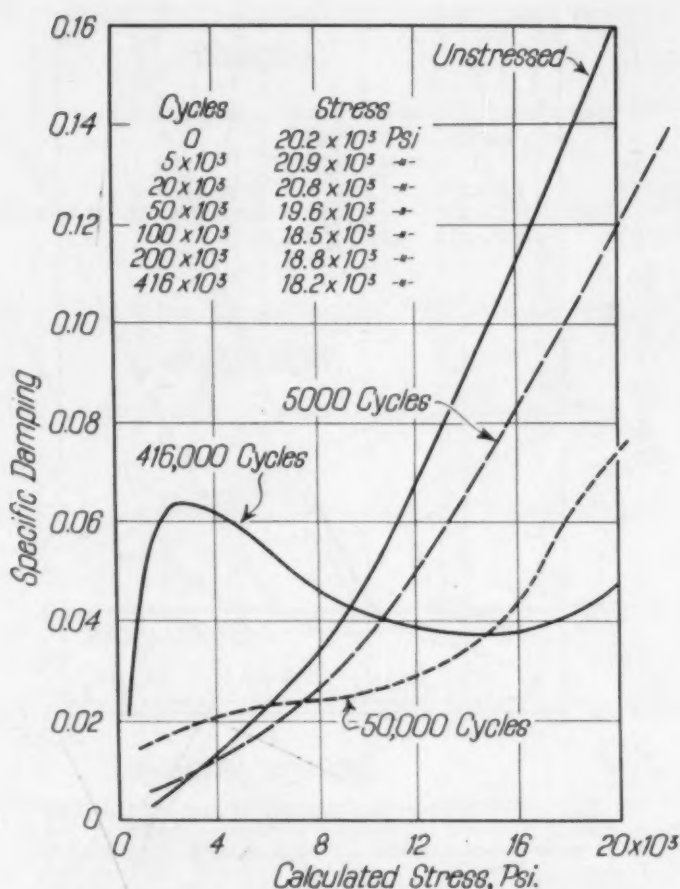
Fig. 13—Changes in Damping of an Alclad 24ST Specimen Stressed in Fatigue Well Above its Endurance Limit. The Corresponding Stress Changes are Tabulated on the Graph.

It is interesting to compare these changes with those accompanying fatigue stressing in the vicinity of, but slightly above, the endurance limit, Fig. 14. It is again evident that the damping capacity in the higher stress ranges sharply decreases while a slight increase occurs in the low range of stress (cf. Fig. 11). The table accompanying the diagram shows that there was no decrease in stress during the constant-deflection fatigue testing while microscopic observations of a polished cross section disclosed evidences of deformation in the aluminum surface layer but no readily detectable cracks.

On the same graph, Fig. 14, the plotted damping capacity of 24ST is shown to be exceedingly low as compared to the Alclad 24ST and by comparison with Fig. 15, less than that of any of the stainless steels. Fatigue stressing of 24ST in the proximity of its endurance limit did not noticeably affect its damping capacity although a slight
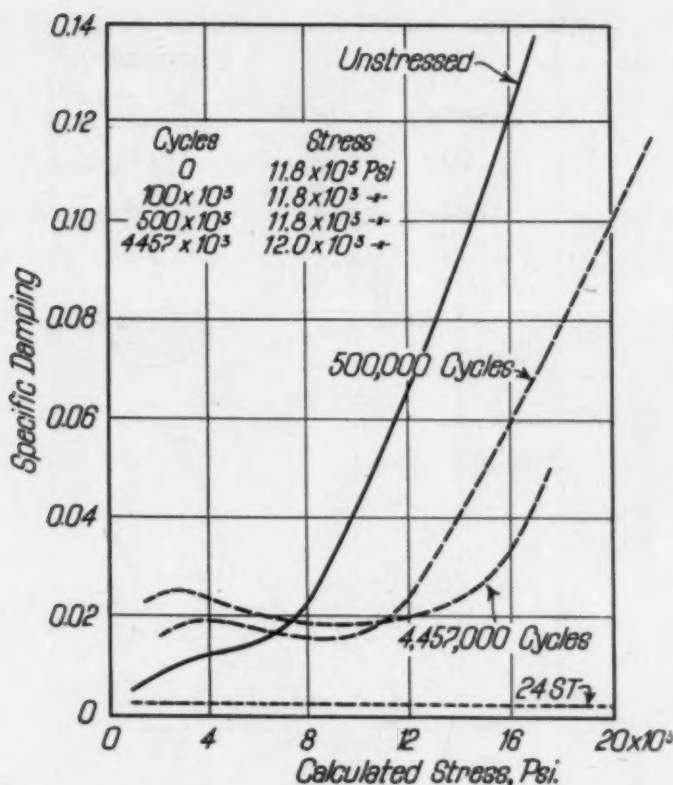
Fig. 14—Changes in Damping of an Alclad 24ST
Specimen Stressed in Fatigue at Close Proximity to its
Endurance Limit. Stress Changes During the Constant-
Deflection Test are Tabulated and in Addition, the
Damping Curve for 24ST Duralumin is Reproduced for
Comparison.

increase in fatigue stress was noted. A number of 24ST fatigue
specimens were polished metallographically without disclosing evi-
dence of cracks.

Damping curves were taken of most of the nine different grades
or conditions of 18-8 stainless steel in their original condition and
the derived damping capacities are reproduced in Fig. 15b. Fig. 15a
shows the effect of repetition of the damping test upon the resultant
curve for three of the steels. Figs. 16 and 17 show the effect of
fatigue stressing below the endurance limit upon the damping capacity
of four different specimens. Examination of the graphs leads to
the following conclusions:

1. Increase in carbon content from 0.07 to 0.11 per cent consid-
erably decreased the damping capacity of annealed 18-8 (cf. Nos. 3
and 4, Fig. 15b).

2. Cold work greatly decreased the damping regardless of car-
bon content (cf. Nos. 3 and 4 with Nos. 5 and 8, Fig. 15b).
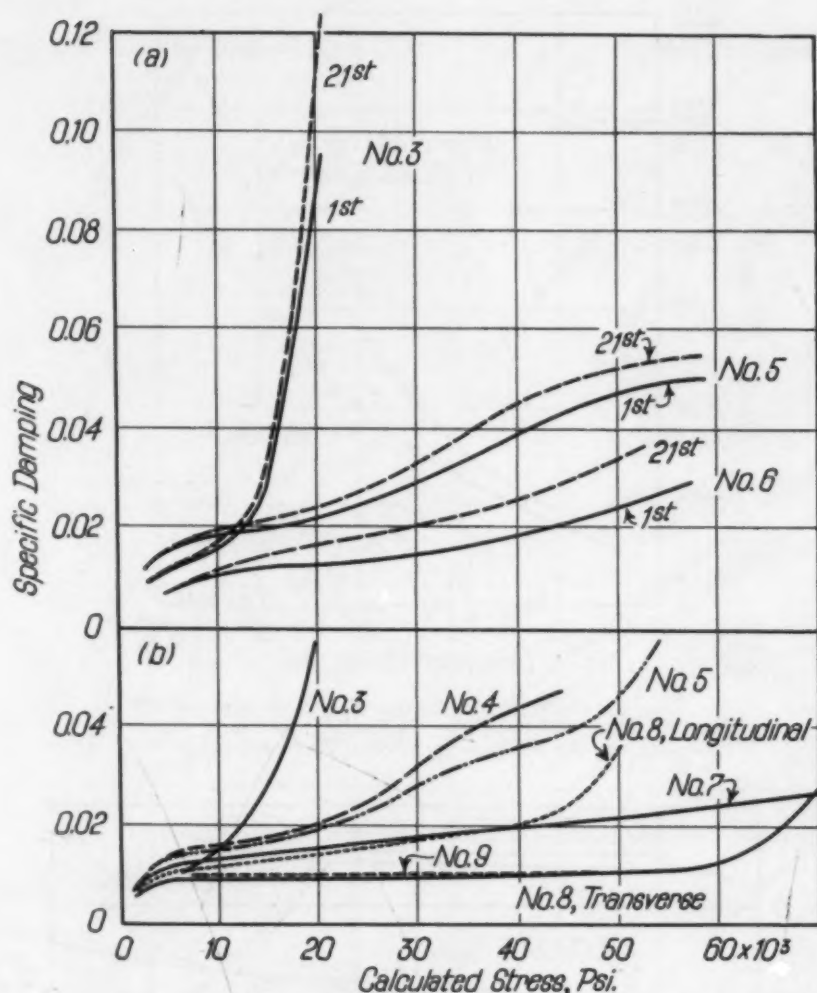
Fig. 15—(a) Changes in Damping of Stainless Steels on Repetition of the Damping Test; (b) Damping Curves for Several of the Stainless Steels: No. 3: Type 304, Annealed; No. 4: Type 302, Annealed; No. 5: Type 304, Cold-rolled; No. 6: Type 302, Cold-rolled; No. 7: Type 347 Cb, Rolled and Aged; No. 8: Type 302, Severely Cold-rolled; No. 9: Special Age Hardened Type.

3. The greater strength and stiffness of transverse specimens of severely cold-rolled 18-8 were associated with a lesser damping capacity (cf. Nos. 8L and 8T, Fig. 15b).

4. There was little difference, except perhaps at high stresses, in the damping of cold-rolled 0.13 per cent carbon stainless steel and the 0.06 carbon columbium-stabilized, age hardened 18-8 (cf. Nos. 7 and 8L, Fig. 15b).

5. The two strongest stainless steels, high carbon, cold-rolled transverse specimens and the special age hardened alloy both showed the lowest damping capacities (cf. Nos. 8T and 9, Fig. 15b).

6. Repetition of the damping test (with the maximum stress
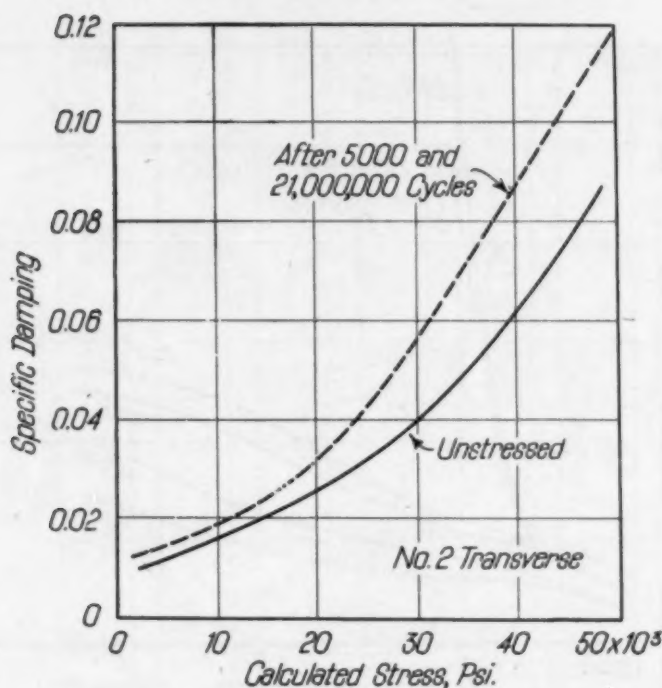
Fig. 16—Change in Damping of Stainless Steel
No. 2 (Type 304, slightly Cold-rolled) After Fatigue
Stressing in the Proximity of the Endurance Limit.
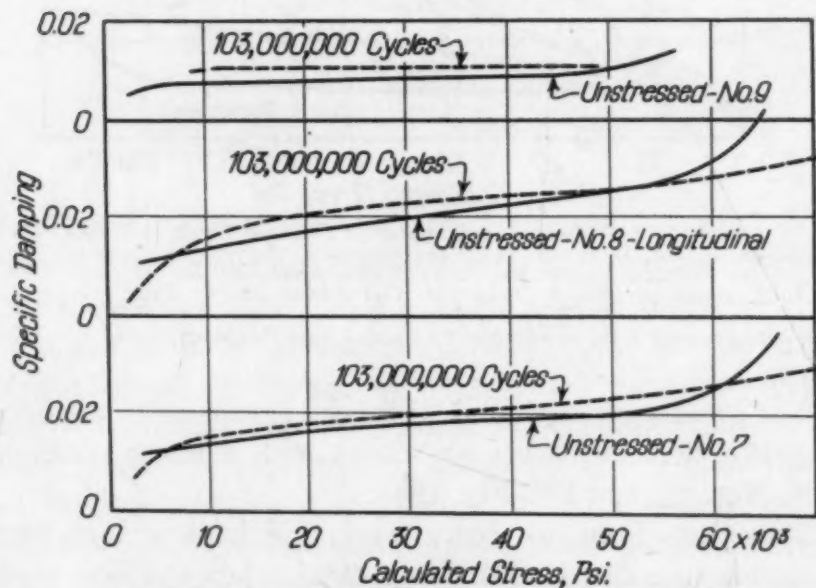
Fig. 17—Changes in Damping of Stainless Steels After Fatigue
Stressing in the Proximity of the Endurance Limit. No. 7 Type
347 Cb, Rolled and Aged; No. 8 Type 302, Severely Cold-rolled;
No. 9 Special Age Hardening Alloy.

above the endurance limit) somewhat increased the damping, par-
ticularly at higher stresses (Fig. 15a).

7. Long continued stressing which started below the endurance
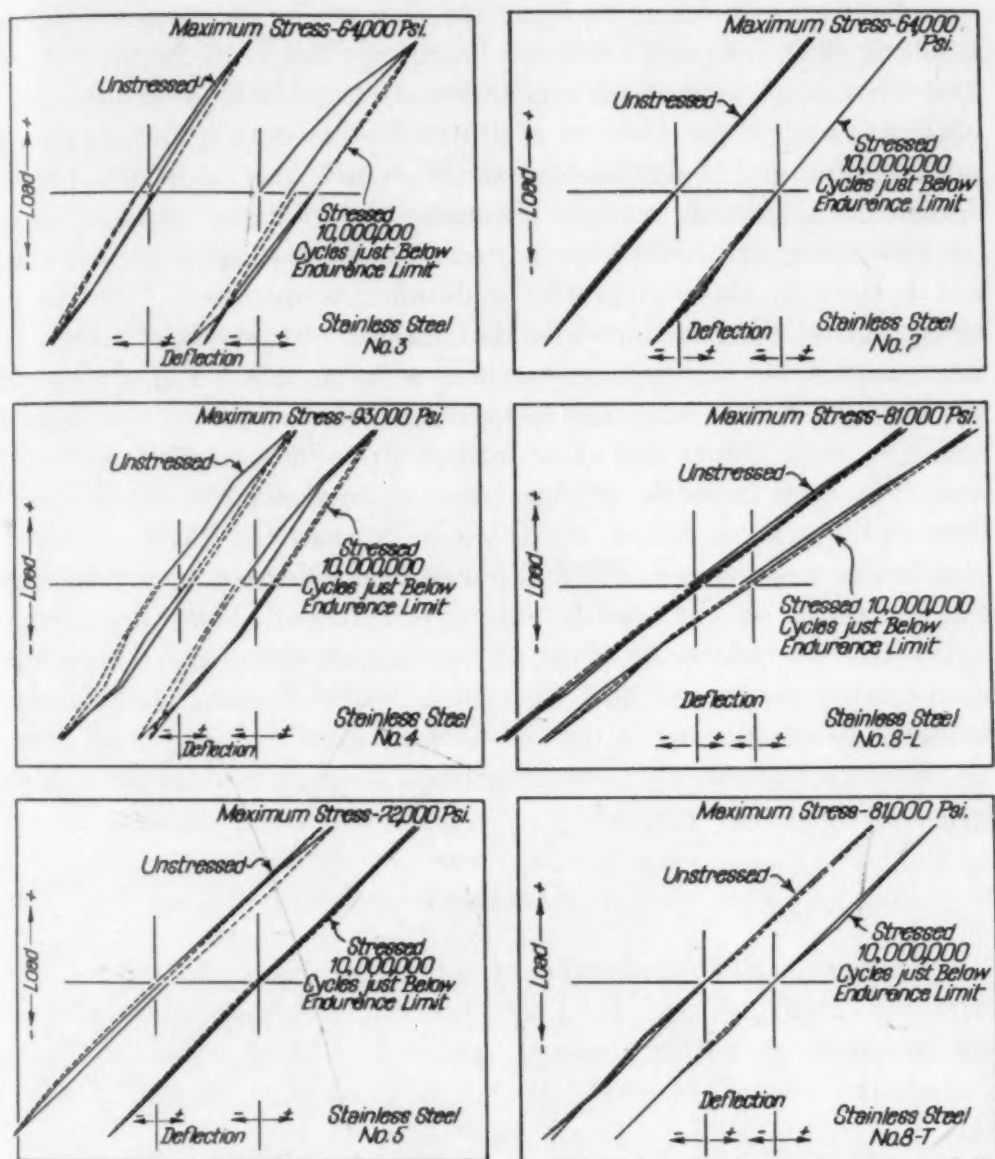
Fig. 18—Mechanical Hysteresis Loops from Load-bending Tests of Stainless Steels in an Unstressed Condition and After 10,000,000 Cycles at Just Below the Endurance Limit.

limit and was continued to stresses above the limit quite noticeably raised the damping of annealed or slightly worked stainless (Fig. 16) and slightly increased the damping of stronger specimens except perhaps at very high and very low stresses (Fig. 17).

8. The very slight increase in damping was probably not associated with the presence of cracks which seemed to be absent in the metallographic sections but it may have been associated with the work hardening accompanying under-stressing.

Another method of studying the changes in stainless specimens resulting from fatigue stressing is based on changes in the mechanical hysteresis loop formed by a complete cycle of a load-bending test. McAdam and Mebs (18) in a study of stress-strain and stress-set relationships in 18-8 stainless steels found that a slight plastic deformation, such as may be encountered in fatigue stressing near the endurance limit, will have a pronounced effect upon the mechanical hysteresis, elastic modulus and other properties. Changes in stress during the constant-deflection fatigue test would naturally be accompanied by a change in mechanical hysteresis. Fig. 18 shows the changes found in a double cycle load-deflection test of several stainless steels before and after fatigue stressing near the endurance limit. In most cases the changes were quite small, considerably less than in the case of Alclad 24ST but more than for 24ST. Several months elapsed between the endurance tests of these specimens and the repetition of the load-bending cycle tests. This time factor would tend to reduce the effect of the fatigue stressing. The maximum stress values on some of these load-deflection curves, particularly those showing a considerable plastic "set", are inaccurate, as mentioned previously, since the beam formula should be applied only in the elastic range.

Summary

1. Approximate endurance limits have been determined for duralumin 24ST, Alclad 24ST and several 18-8 stainless steels with low and moderate carbon contents, annealed and cold-rolled, stabilized and age hardened. The results are tabulated on page 441 and in Table III.

2. The relatively low fatigue strength of Alclad 24ST seemed to be related to the formation of large numbers of shear cracks in the weak aluminum surface layer and their propagation into the alloy core.

3. Crack formation in Alclad resulted in a decrease in the calculated maximum stress during the constant deflection fatigue test. The 24ST and most of the stainless steels usually showed an increase in the apparent stress, presumably as a result of work hardening from plastic behavior through the stress cycle.

4. Despite a considerable scatter of points on the S-N fatigue curves of the stainless steels, the data showed a general tendency

for the endurance limit to remain in the range of 30 to 40 per cent of the tensile strength for the commercial finish No. 2-B, a bright finish which, nevertheless, shows etched grain boundaries from the prior pickling operation. Severe cold rolling after pickling did not remove the grain boundary etch effect nor did it appreciably raise the endurance ratio. An extremely fine ground finish, No. 4, obliterated the etched boundaries and increased the endurance limit to the range 60 to 70 per cent of the tensile strength.

5. The scatter of fatigue data may be attributed in part to an incomplete knowledge of stress-changes during the test and in part to mechanical errors in setting up the machine and stresses in the sheet but in the case of the stainless steels, the variability is probably related to work hardening in the highest stressed surface layers versus the possibility of a minute surface defect originating a crack.

6. Fatigue stressing of Alclad 24ST above the endurance limit sharply decreased the damping at high stresses and as failure became imminent, i.e., fatigue cracks were propagated into the core, the damping at low stresses greatly increased. Fatigue stressing of Alclad in the vicinity of the endurance limit showed the same sharp decrease in damping at high stresses but not the later increase at low stresses.

7. Among the stainless steels, the damping capacity decreased almost continuously with increase in tensile strength. The details of the changes with carbon content, cold rolling, aging and fatigue stressing have been listed on pages 456 to 459.

8. Mechanical-hysteresis loops or complete load-bending cycles of fatigue specimens gave a qualitative idea as to relative damping capacity but actual damping curves were required for adequate and quantitative definitions of slight changes.

Acknowledgment

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DISCUSSION

Written Discussion: By F. J. Dawless, manager, Crucible Steel Company of America, New Haven, Conn.

I certainly wish to congratulate the authors on their far-seeing paper covering a subject that is of vital importance in connection with our National defense program. Surely, the question of the use of all-stainless airplanes will be more and more in the limelight as our aircraft industry turns toward mass production.

There are a few points I should like to question. I am, of course, dealing only with the portions of the paper on stainless steels, since I do not consider myself sufficiently informed to comment on the work on the aluminum alloy 24ST and Alclad 24ST.

The fatigue tests reported by the authors confirm the results obtained by a number of people, including Faulhaber and Moore, on the dominant effect of surface on fatigue strength. I am rather skeptical, though, of the very high ratio of endurance limit to tensile strength found in the two samples with the No. 4 finish. It would have been much more convincing to me if two of the other samples with the No. 2-B finish, for example, Nos. 3 and 4, had been given a No. 4 finish and then retested to see if finish was really the only difference. I would also have liked to see fatgiue tests made on polished samples of several of these steels, as I think that these results would indicate theoretical maximum endurance limits that would be more representative of the material than the results on the as-rolled finishes. The authors mention that they hoped that the latter would be more representative, but it is my feeling that any as-rolled surface contains too many mechanical imperfections to be used in tests of this kind, since the results will be more typical of the finish than of the material itself.

I am very much interested in Fig. 8b, which, the authors consider, indicates that the grain boundaries of the steel have been deeply attacked during the final pickling operation. Frankly, I am not used to such photomicrographs of the surface of stainless steel, and would not wish to reach any conclusion on the results obtained until I had seen photomicrographs prepared in the usual manner. I am not at all convinced that Fig. 8b does represent intergranular attack, and would like to see data to prove this.

The authors, in Conclusion 1 on the results of the damping tests on stainless, attribute the difference between steels 3 and 4 (in Fig. 15b) to the difference in carbon content of the steels tested. I wonder why they feel that it is due to carbon content rather than to the difference of 14,000 pounds per square inch in the tensile strength of these two grades. From the results given, it is noted that there is a fairly clear relationship between hardness, or tensile strength, and damping. I therefore wonder why this is considered insufficient to explain the greater damping capacity of steel No. 3.

Incidentally, I note that the results on Steels Nos. 3 and 5, as given in Figs. 15a and 15b, do not quite check. Furthermore, the shape of the curve for steel No. 5 in the two figures is not the same. I wonder to what the authors attribute this difference.

The general consensus of opinion has always been that heterogeneous materials and structures had greater damping capacity than did homogeneous ones. I would therefore have expected that the 18-8 Cb would have a greater damping capacity than 18-8 because of the presence of undissolved columbium carbides and other particles always present in the columbium type. It was rather amazing, therefore, to see that steel No. 7 had, if anything, lower damping capacity than the longitudinal specimen of steel No. 8 at practically the same tensile strength.

In Conclusion 8 on the damping capacity of stainless steel, the authors state that cracks seemed to be absent in the metallographic sections. In cases such as those under consideration, where the presence or absence of cracks is of considerable importance, it is of interest to know just how carefully and by what means the authors investigated the samples to see if cracks were or were not present. A few random metallographic sections would not seem to me to be adequate.

Have the authors checked the damping figures obtained on their damping machine with any other damping machine? From the results that I have seen on damping capacity, there is very often an appreciable difference that can be attributed to the construction of the machine.

I think the authors have made a splendid start on this difficult subject, and I would be glad to see further work from them on the effect of corrosion on the fatigue limits of these materials. Although the corrosive action of the atmosphere, sea water, etc., would not be great on the stainless steels under consideration, yet McAdam's work has indicated the possibility that even this very slight corrosion might change the fatigue strength. I feel sure that such a study would be of great interest and assistance to designers and steel makers.

Written Discussion: By N. B. Pilling, research laboratories, International Nickel Co., Bayonne, N. J.

The interesting comparisons the authors make regarding the fatigue habits of two important types of sheet metal for aircraft bring to mind one point on which the paper is not quite clear. One is led to assume that even after severe and prolonged cyclic straining the stainless test specimens were quite free from penetrating shear cracks. It would be appreciated if the authors would take a more definite position on this and include micrographic sections to reveal what, if any, structural alterations occur at and below the surface of the stainless steels with which they worked.

Oral Discussion

E. H. Dix, Jr.:[1] It is sometimes fortunate that a material gets into use before we begin testing it too far. One of the biggest problems that confronts us is the interpretation of test results. For those of you who may not be entirely familiar with the airplane industry, I should like to say that there has been a progressive increase in the use of Alclad 24ST as replacing bare 24ST over a period of ten years.

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[1]Chief metallurgist, Aluminum Research Laboratories, Aluminum Company of America, New Kensington, Pa.

The point I want to make is that somewhere there is a discrepancy between practical use and interpretation of such data as Dr. Brick has shown us.

I do not know whether I can help to indicate some of these differences; there are some questions I would like to ask. Unfortunately, to run a fatigue test on a sheet material using a simple sheet specimen, about the only thing you can do is to bend it back and forth. Of course that puts the surface in the place of maximum stress as farthest away from the neutral axis. No aircraft designer would use a piece of sheet under those conditions, if he wanted to carry the maximum load. In other words, Alclad parts are generally so designed in channel sections, et cetera, so that the whole sheet is at considerable distance from the neutral axis, rather than just the surface. In other words if it were possible to make a satisfactory tension fatigue test on an Alclad sheet, undoubtedly the answer would be different from the one you get when you bend it back and forth. Then comes the question as to how these stresses are calculated.

It would be one thing to assume that the pure aluminum coating is purely a paint protection, and use the thickness of the core for the stress calculations; it is another thing to use the full thickness of the sheet and assume, when we know it is not so, that the stress bears a straight line relationship from the surface to the neutral axis.

This is not the first time where aluminum alloys have gotten into service before fatigue test data were available. One alloy that I have in mind is a heat treated casting alloy used extensively for those applications where shock and fatigue resistance are required, and yet on smooth specimens tested in fatigue—that alloy has a lower endurance limit than the one that it successfully replaced.

In regard to damping capacity, there are several things that I think may be of interest. At one time the Bureau of Standards attempted to run fatigue tests on the Alclad sheet by a method in which they set up vibration by a stream of air; they were able to make the tests on 24ST quite satisfactorily, but they could not get the Alclad material to vibrate, which again may have some bearing on the practical usefulness of it.

Again, some ten years ago a metal-clad dirigible was built in which Alclad sheet 0.0095 inch thick was used for the hull and served without a lining for the gas bag; that was a beautiful place for vibration to be set up, but the hull did not vibrate.

I do not want anything that I have said to detract in any way from the very beautiful piece of work that Dr. Brick and his associates have done; at the same time I did not like to leave the impression that the material that the aircraft people are using to an ever increasing extent has an endurance limit in service 40 per cent lower than the material that it is replacing.

V. N. KRIVOBOK:[2]   I have not yet seen any figure as low as 22,000 pounds per square inch fatigue strength for stainless alloys. I have not, in all the work that we have done, encountered as low a figure as 30 per cent of the tensile strength. Furthermore I would like to point out one thing: In studying

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[2]Formerly with Allegheny-Ludlum Steel Corp., Brackenridge, Pa.; now director of structural research, Lockheed Aircraft Co., Burbank, Calif.

the tables and values of the figures presented in the paper, I notice quite a discrepancy between comparable columns. In Table II, where the so-called test values and the nominal values are given; for example, the yield point (which is a misnomer, it should not be called a yield point, it is a yield standard) is given as 200,000 pounds and yet as designated by the authors, gives a value of 135,000 pounds per square inch. That is a very serious discrepancy which has to be explained.

We all know that the elastic properties of 18 K, especially under these conditions, are very low; if no care is taken to see to it that the initial load—and I mean first initial load—is sufficiently low, the sheet which you are testing does not come up to its original position, hence it has taken on a certain amount of permanent set, and as Dr. Dix pointed out, it is immediately cold-worked on the surface, and therefore you are measuring something which is changing as your test progresses.

We have found out that unless we are very careful in keeping our tests under the load which would correspond to the so-called yield strength, our values will vary widely. If, however, we take care of that, we have not yet found a single composition, nor cold rolling condition under which the 18-8 stainless alloys will not show fatigue strength considerably higher than has been indicated.

I agree with the remarks of Mr. Dawless and suggest that a technique for etching stainless steel may be considerably improved. The photomicrographs shown in the paper do not represent the material that you will usually obtain.

Now as to the damping capacity: I believe that in the field of damping capacity, the effect of the composition, the method of testing, etc., are not sufficiently well known as yet to permit more careful consideration. I do agree with the authors and disagree with Dr. Dix on this, that there is nevertheless a good safety factor put in on top, so we indeed welcome your contribution, because it is on the basis of work such as yours that the engineers are basing their designs and maybe, if our values are better substantiated, then the engineers will reduce their factors and build better and faster airplanes.

Oral Reply

R. M. BRICK: Details given in the paper do take care of some of the points brought up by Messrs. Dix and Krivobok. Although there was not sufficient time in the presentation to discuss these, it might be appropriate to review certain details now.

For example, in regard to the discrepancy in the nominal yield points and the test values as given in Table II; the nominal yield point was that scribbled on the sheet, as received directly from the producer. We have no knowledge of the basis for that yield value. The actual value given as a result of our tests (of the stainless steels) was the value at a 0.10 per cent offset from the straight stress-strain line.

In regard to the questions as to the actual stress values (this also an-

swers the point as to stainless steel with a yield point of 26,000 pounds per square inch) ; as you all know, stainless steel has an unfortunate habit of deforming plastically at stresses far under the endurance level. The only stress formula the authors have knowledge of is the ordinary stress formula which presumes purely elastic action. It is stated in the paper that all our stress values are subject to error on that source; that error is greatest for the softest annealed stainless steel, and is, of course, less on the hard-rolled structures.

The same comment applies to the stress values for Alclad which also showed plastic behavior during the stress determination—the weak aluminum surface layer deformed. This leads to an error whose value we cannot readily estimate.

Both Mr. Dawless and Dr. Krivobok feel that the surfaces as presented in Fig. 8 of the paper do not represent commercial surfaces. The authors cannot answer that other than to say that these were obtained through commercial sources; they were the same sheet, for example, that a man who is going to draw a stainless object, a hub cap, might be buying.

As to whether the photographs represent the surface truly or not: The commercial sheet, as it was received, was put on the microscope at 200 diameters and photographed. The surfaces were not questioned by representatives of the producing company when they viewed the photographs. We cannot say definitely that these pictures (Fig. 8) are typical of commercial finishes, but have every reason to believe they are.

In answer to Dr. Dix, we were somewhat reluctant to publish our results for fear that the test data might be interpreted as indicative that Alclad should not be used. We were very careful to avoid any intimation that Alclad was an unsafe material for construction in aircraft use. In the paper it is emphasized that the high damping capacity of the Alclad will automatically serve to protect it from excessive vibration. Föppl and other workers in damping have emphasized that a material with a lesser strength is frequently far more serviceable for certain applications if it has a high damping capacity; the high damping capacity will prevent the system from starting to vibrate or from vibrating at dangerously large amplitudes. In addition to the safety factor that Dr. Krivobok mentioned, that is undoubtedly one of the reasons why Alclad has been so successful in aircraft use.

In reply to Mr. Pilling and also to one of Mr. Dawless' comments as to cracks in the stainless steel, the authors must admit that they did not examine many stainless steel specimens for shear cracks and it is possible that an examination of all specimens would have disclosed cracking, particularly in steels with the 2B finish.

The question has been raised as to the low damping capacity of columbium stainless steel which seems to be contrary to the generally increased damping of a heterogeneous structure. The effect of a second phase depends on its dispersion; cast iron has a high damping capacity but, as shown in studies of aluminum-copper alloys, damping decreases with age hardening and the low damping capacity of the columbium stainless steel may be explained on the basis of a fine carbide dispersion.

Authors' Written Reply

Mr. Pilling's request for a clearer statement regarding the susceptibility of the stainless steels to shear cracking led the authors to polish sections from about fifty more stainless fatigue specimens in addition to a few previously examined metallographically. Only two specimens showed shear cracks and in each case there was a single crack. One is led to the same assumption as that tentatively stated by Mr. Pilling, that stainless steel is relatively free from the shear cracks found so abundantly in Alclad.

In further reply to Dr. Krivobok, the authors are at a loss to understand how fatigue tests can be run at stresses under the yield strength so that no deformation and related strain hardening can occur, if the fatigue strength is above the yield strength. It seems to be generally agreed by all experts in the field of fatigue testing that the endurance limit of most metals is above their proportional limit; some plastic deformation can continuously occur with an accompanying rise in specimen temperature and a fragmentation of the crystal structure, and yet with no sign of ultimate failure (up to a billion or more cycles of stress).

The difference in damping of steels Nos. 3 and 4 is attributed to carbon content rather than strength, as suggested by Mr. Dawless, since the increased strength of No. 4 is explicable on the basis of its higher carbon content. Differences in the damping curves of steels Nos. 3 and 5 in Figs. 15a and 15b are relatively minor and may be explained on the basis of errors in the determination of damping capacity at the higher stresses. In general, it may be said that *relative* damping capacities as affected by fatigue stressing were the data of interest in this work and absolute values for damping were considered unimportant. However, Föppl has shown that damping capacities determined by different methods usually agree well with one another and the numerical damping data given here agree moderately well with data obtained by entirely different methods.

The authors agree with Mr. Dawless that it would be desirable to take some of the present material having a No. 2B finish with a determined endurance-tensile strength ratio of 30 to 40 per cent; giving this material a No. 4 (ground) finish and retesting to see if the ratio was increased to 60 to 70 per cent. However, specimens Nos. 1 and 3 (Table I) should be nearly identical except for the additional grinding given the surface of No. 1. As Mr. Dawless says, the data represent surface finishes rather than the ultimate metal and that point is one stressed in the paper.

Dr. Krivobok and Mr. Dawless doubted that the photomicrographs of Fig. 8 portray the surface conditions of commercial stainless steel sheets having finishes conforming to standard practices. If they have been sufficiently interested in this question to examine the surface of commercial material we are sure that they are now ready to withdraw their objections to our findings. Since the presentation of our paper we have seen abundant evidence, from the files of a stainless steel producer, showing without question that our photomicrographs are quite typical of present day surface finishes. We are including

in this discussion two photomicrographs, Figs. 19 and 20, from this source which are freely conceded to be representative of surface conditions of the sheets produced by this mill. Without doubt the surface pattern represents the combined effects of oxidation and pickling during processing. We regret
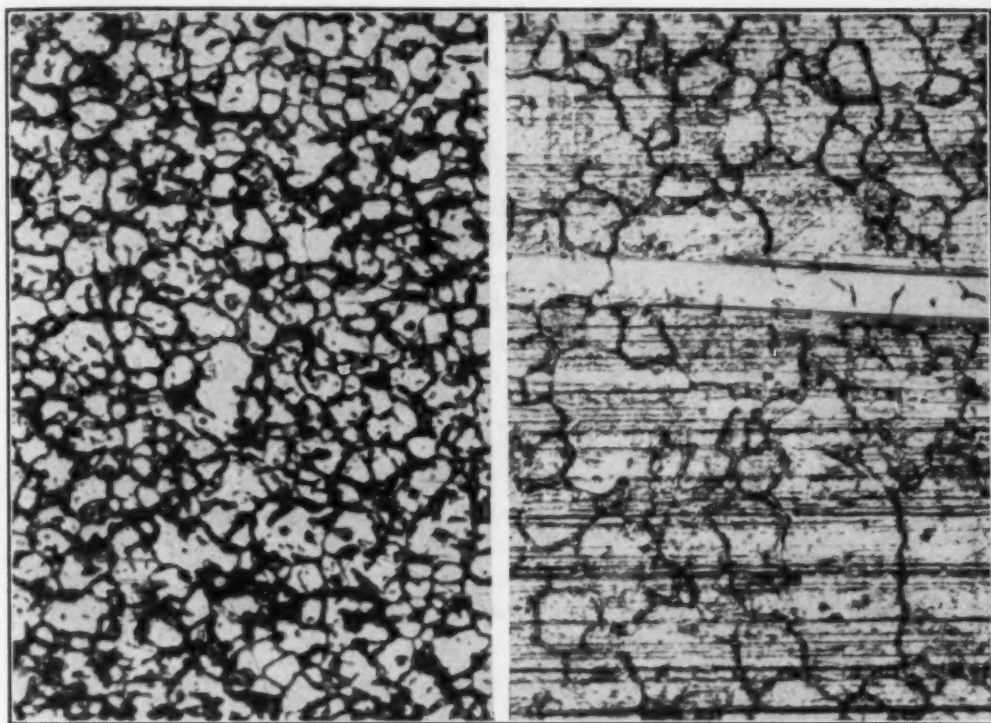
Fig. 19.   Fig. 20.

Fig.-19—Type 302, 18-8 Stainless Steel; Commercial Sheet with a No. 2B Finish. × 200.
Fig. 20—Type 302, 18-8 Stainless Steel; Commercial Sheet with a No. 7 Finish, Muffle Annealed and Pickled. × 200.

overlooking an important paper by Baeyertz[3] who showed that oxidation of 18-8 stainless at 2000 degrees Fahr. is preferentially intergranular. The network shown in our photomicrographs may be attributed to the removal of oxide by the pickling acid.

[3]TRANSACTIONS, American Society for Metals, Vol. 24, 1936, p. 420.

SOME SURFACE STUDIES ON TREATED HIGH SPEED STEEL

By J. G. Morrison

Abstract

The probable disposition of the carbon (as regards percentage) has been determined in the zone approximately 0.001 inch below the surface to about 0.012 inch below the surface of 0.60, 0.70 and 0.93 per cent carbon 18-4-1 high speed steels treated at 2350 degrees Fahr. in two widely divergent atmospheres, viz., one containing 10 per cent carbon monoxide and one containing 2.5 per cent oxygen. Using a constant holding time in the 2350 degrees Fahr. furnace the evidence indicates a greater absorption of carbon when the 2.5 per cent oxygen containing atmosphere is used as compared to the 10 per cent carbon monoxide containing atmosphere. The 0.93 per cent carbon high speed steel absorbed carbon at the surface to approximately 0.95 per cent in either atmosphere.

Oxidation or loss of surface by scaling in the 2350 degrees Fahr. furnace appears to be dependent, to some extent, on how rapidly the surface attains to approximately 0.95 per cent carbon.

Microscopic and microhardness (Knoop) data confirm the analytical data of the "surface" zones of the various treated 18-4-1 high speed steels.

The 0.60, 0.70 and 0.93 per cent carbon 18-4-1 high speed steels when coated with borax prior to heating at 2350 degrees Fahr. and molybdenum high speed steels heated for hardening (with or without a borax coating) at lower temperatures in various atmospheres all show some carbon absorption at the surface under the conditions imposed.

THERE appears to be no reference in the literature of any study having been made of the disposition of the carbon (as regards percentage) in the extremely important zone from the treated surface of 18-4-1 high speed steel to approximately 0.012 inch below the surface. The purpose of this investigation was to find out more about this zone since it constitutes the edges of many cutting tools

and consequently has an important bearing on their performance.

For some time a number of the larger tool manufacturers have been aware that there is considerable chemical alteration (as regards carbon) of the surface and immediate subsurface of 18-4-1 high speed steel which has been given a normal heat treatment. A few years ago (1936) Mr. J. P. Gill, Chief Metallurgist of the Vanadium Alloys Steel Co., and the author prepared a number of $\frac{15}{16}$-inch diameter by 5-inch long samples made from a bar of 0.72 per cent carbon 18-4-1 high speed steel and subjected them to three different treatments as indicated in Table I. The treated samples were cleaned by pickling and surface cuts were made on each sample without any annealing treatment being used. A hard carbide tool

Table I
Carbon Determinations on "Surface" Cuts of 0.72 Per Cent Carbon 18-4-1 High Speed Steel in the As Quenched Condition

Sample No.	Preheat 1550 Deg. Fahr.	High Heat 2350 Deg. Fahr.	Quench	Carbon Determinations 0.000 to 0.003 Inch—0.000 to 005 Inch		
				Combined Carbon	Graphitic Carbon	Total Carbon
2	22 Min.	None	Oil	0.72	0.07	....
4	None	5½ Min.	Oil	....	....	0.96
1	22 Min.	4½ Min.	Oil	0.86	0.065	....

was used for securing the chips. It was observed in machining one of the samples (sample No. 4) that the chips were quite sooty. Preliminary carbon determinations on this sample showed the carbon on the surface cut 0.000 to 0.005 inch to be 0.96 per cent. Consequently after machining sample No. 1 an additional determination was made for graphitic carbon. Table I shows the results obtained as regards the carbon contents of the surface layer of each of the three samples. The preheat atmosphere contained 6 per cent CO and the 2350 degrees Fahr. (1285 degrees Cent.) atmosphere contained 9 per cent CO and 13 per cent CO_2.

It was believed the graphitic carbon found was due to a soot deposited either from the furnace atmosphere or from the cracked oil in quenching. The results obtained in Table I might indicate a soot may deposit on the steel surface in the preheat but no carburization takes place; carburization does occur in the high heat. That a deposited carbon may be difficult to remove entirely on pickling may be due to the nature of the treated surface as influenced by the atmosphere and the high temperature employed.
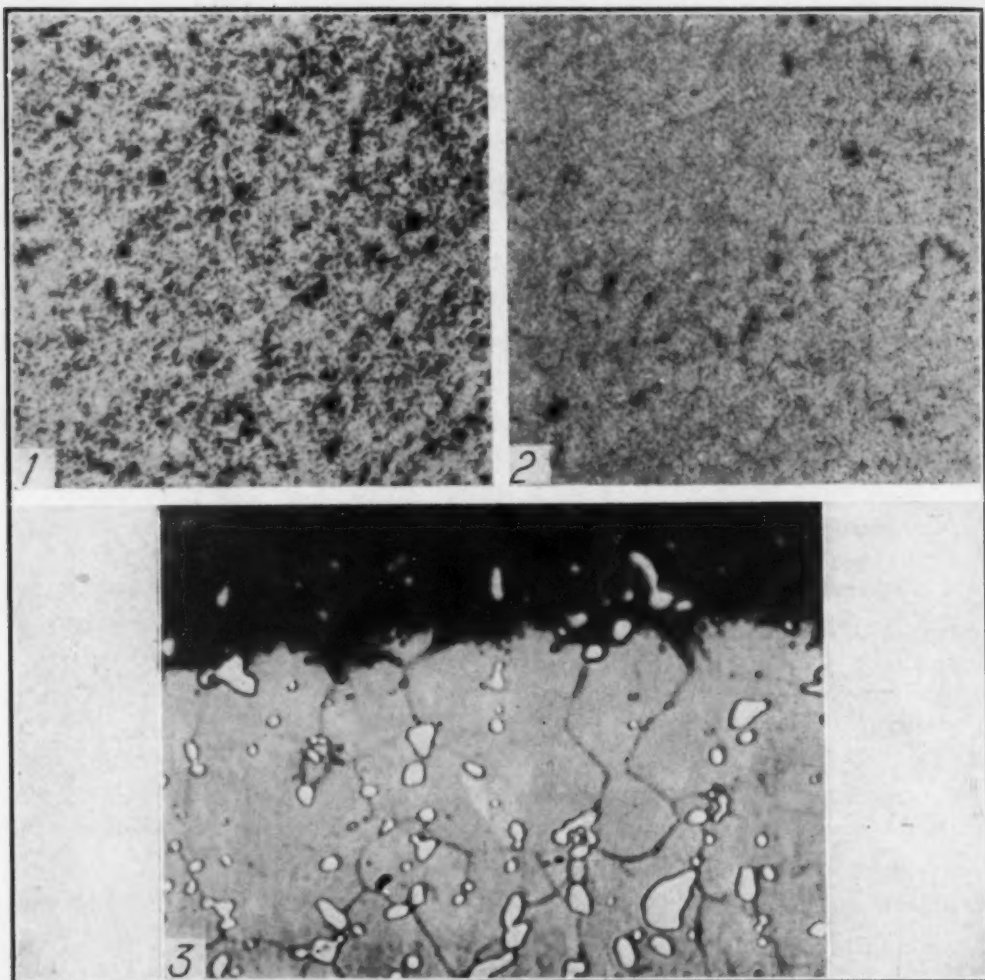
Fig. 1—Surface of a 0.70 Per Cent Carbon 18-4-1 High Speed Steel Polished Before Hardening at 2350 Degrees Fahr. in an Atmosphere of 10 Per Cent CO. Surface Repolished Lightly with "0000" Paper and Finer After Oil Quenching. Unetched. × 200.

Fig. 2—Same as Fig. 1 Except Sample Was Quenched Into a Salt Bath Held at 1100 Degrees Fahr., Finish Cooled in Air. Unetched. × 200.

Fig. 3—Section of a Milled Surface of a 0.75 Per Cent Carbon 18-4-1 High Speed Steel Hardened from 2350 Degrees Fahr. Using an Atmosphere of 10 Per Cent CO. As Quenched Condition. Etched Nital. × 1000.

If one polishes a high speed steel specimen and hardens it from 2350 degrees Fahr. (1285 degrees Cent.) and then repolishes it lightly using "0000" paper and finer it will be found to contain minute pits and a pattern suggesting that the surface attack is most avid at the grain boundaries. Figs. 1 and 2 show the surfaces of the 18-4-1 steel specimens so prepared and micrographed unetched at × 200. If extraneous carbon were to be lodged in the small pits shown or entrapped in the attendant scale it is possible that pickling alone might not entirely dislodge all surface adhering carbon, par-

ticularly so, if the acid pickle did not free the surface from all scale.

Fig. 3, a photomicrograph taken at $\times$ 1000 of a section of the hardened (2350 degrees Fahr.) (1285 degrees Cent.) undrawn surface of a 0.75 per cent carbon 18-4-1 high speed steel, illustrates the attack at the grain boundries above mentioned. Undecomposed carbides may be observed in the outer zone of scaled metal.

Observations covering a number of years have indicated that there is no apparent difference as regards the tendency of 18-4-1 steel to decarburize in rich or lean CO atmospheres (14 to 1.5 per cent CO) provided a reasonable time cycle is employed on samples of moderate size. On the contrary with fast heating rates and relatively short time cycles in the 2350 degrees Fahr. furnace, using samples with sharp projections, there is microscopic evidence of a more extensive carbon absorption when the lean CO atmosphere is used compared with the rich CO atmosphere.

The original purpose of this investigation was to compare the surface carbon of 18-4-1 high speed steels hardened from 2350 degrees Fahr. using (a) an oil fired semi-muffle furnace in which an atmosphere containing 1.5 per cent CO prevailed and (b) a curtain furnace using an atmosphere containing 10 per cent CO. In a practical research conducted about a year and a half ago by the Carpenter Steel Company and in which the author's company cooperated, they found that 18-4-1 high speed steel evidently absorbed carbon in O_2 atmospheres at an even greater rate than in the CO atmospheres. The Carpenter Steel Company was probably the first to recognize the greater tendency for carburization by the treated surfaces of 18-4-1 high speed steel when atmospheres containing some oxygen are used compared to CO atmospheres. With this knowledge our original experiment was modified so as to use an atmosphere containing a definite amount of oxygen instead of the 1.5 per cent CO atmosphere. The 2.5 per cent O_2 atmosphere was selected as it was known that a least one company had used such a controlled atmosphere. The selection of the 10 per cent CO and 2.5 per cent O_2 atmospheres probably represents the extremes of carbonaceous atmospheres between which nearly all 18-4-1 high speed steels are treated. The same controlled atmosphere furnaces were used for all the tests.

The chemical changes taking place on the surface and immediate subsurface during the sojourn in the 2350 degrees Fahr. furnace are discussed here with respect to a round bar and are limited to a

consideration of the changes of the carbon content only. The altera-
tion of the surface chemistry of 18-4-1 high speed steel during treat-
ment is complicated by a number of factors and the disposition of
the carbon is by no means uniform on all surfaces of a treated piece.
The disposition of the carbon for approximately the first 0.012 inch
of "surface" is markedly affected by the shape of the piece and we
would find the carbon to vary on a round surface (or a flat surface)
from that found on a corner or on the sides of an acute re-entrant
angle. A round bar approximately ⅞ inch in diameter was selected
for the study of the carbon in the surface zone because it is a shape
most conveniently handled as regards measurement, straightening,
and for securing the chips for chemical analysis.

The steels involved in the tests for determining the disposition
of the carbon in the zone from the surface to about 0.012 inch below
the surface of treated high speed steel are given in Table II.

Table II
Various Heats of High Speed Steel Involved in the Tests for "Surface" Carbon

Heat	C	Mn	P	S	Si	Cr	W	V	Mo	Ni	Co	Fe
"Y"	0.60	0.20	0.016	tr	0.35	4.25	18.28	1.03	0.31	0.061	0.07	74.98
"X"	0.70	0.21	0.021	tr	0.28	4.00	18.00	1.01	0.21	0.078	0.02	75.22
"Z"	0.93	0.26	.....	.....	0.20	4.17	17.88	1.13	0.13	0.110	0.00	74.27
"W"	0.71	0.21	0.018	tr	0.33	3.98	18.16	1.03	....	0.06	....	.....
"V"	0.80	0.27	0.022	0.014	0.28	4.40	.....	2.15	9.25	.....	....	.....

Table III
Furnace Atmospheres Used in the Tests for the "Surface" Carbon
Determinations and for the Micrographic Specimens

Atmosphere	Temperature Degrees Fahr.	Per Cent		
		CO_2	CO	O_2
"A"	1450	15.2	2.0	0.00
"B"	1550	15.6	2.0	0.00
"C"	2200	11.8	11.7	0.00
"D"	2200	11.6	0.00	2.5
"E"	2350	10.8	10.0	0.00
"F"	2350	12.7	0.00	2.5

The furnaces used for treating the various samples consisted
of an electrically heated preheat furnace equipped with a gas cur-
tain and a curtain furnace equipped with pin type globars (air-cooled
terminals).

The various atmospheres used in both the preheat and high
heat furnaces are given in Table III.

The atmospheres noted in the tables and charts will be desig-
nated, for the sake of simplicity, as to the amount of CO or O_2

they contained. It is possible that the results obtained in this investigation might not be duplicated in localities where the components of the atmospheres other than those listed were to differ greatly.

Pertinent data in regard to the preparation of the samples for the "surface" carbon determinations are given in Table IV. Sandblast cleaning of the lead "annealed" samples was resorted to, to remove the scale.

Table IV
Preparation of the Samples Used for the "Surface" Carbon Determinations

	Steels			
	"Y"	"X"	"Z"	"V"
Size bar stock	2 2/32" x 3¾"	1" Dia.	1" Dia.	1" Dia.
Length of samples	5"	4" or 5"	5"	5"
Approximate rough turned diameter	0.835"	0.895"	0.895"	0.775"
Approximate diameter turned for securing chips for check for decarburization	0.825"	0.885"	0.885"	0.765"
Approximate diameter after grinding between centers	0.820"	0.880"	0.880"	0.760"
Diameter measured before hardening	1 inch from unstamped end.			
Treated	As indicated in Tables V, VI, VII and XI.			
Position of samples as placed in furnaces for hardening.	Vertical—End stamped with heat symbol and sample number resting on the hearth.			
After drawing	Pickled in cold 25% HCl using an inhibitor. Diameter measured 1" from unstamped end. Distortion measured between centers with dial indicator contacted in middle of sample.			
After lead "annealing", 1425 deg. F.— min. cooled in silocel for steels "Y", "X" and "Z". 1400 deg. F.—10 min. for steel "V".	Cleaned under light sandblast. Diameter measured 1" from unstamped end. Chips for carbon determinations obtained, using Tantalum carbide tools, as indicated in Tables V, VI, VII and XI. Samples turned from unstamped end after a discarded facing of ¹⁄₃₂" to within 1" of stamped end.			

Roughly one half the difference between the treated diameter and the lead "annealed" and sandblast cleaned diameter represents the surface of the treated bar which was sacrificed in the "annealing" and cleaning. No estimate of volume change was attempted as it was believed this would be too small in a 0.880-inch diameter sample to interfere materially.

The present study may be divided into the following parts:

(a) A preliminary inspection of the effect of time (in the 2350 degrees Fahr. furnace using a 10 per cent CO atmosphere) on the carbon content of three successive surface cuts 0.003 inch (0.006 inch on the diameter) on a 0.70 per cent carbon 18-4-1 high speed steel.

(b) The disposition of the carbon from approximately 0.001 inch below the surface of hardened, drawn and pickled 18-4-1 high speed steel to approximately 0.012 inch below the surface. Probability curves of the disposition of the carbon in this zone were developed

for a 0.60 per cent carbon and for a 0.70 per cent carbon 18-4-1 high speed steel hardened from 2350 degrees Fahr. using (1) an atmosphere containing 10 per cent CO and (2) an atmosphere containing 2.5 per cent O_2. A 0.93 per cent carbon 18-4-1 high speed steel was prepared to simulate the carbon content found in the early surface test cuts (Table I). The 0.93 per cent carbon material was treated and tested along with the 0.60 and 0.70 per cent carbon heats.

(c) The effect of the use of borax as a protective coating in heating for hardening on the disposition of the carbon in the zone about 0.0008 inch below the treated surface to about 0.012 inch below the surface. Samples of 0.60, 0.70 and 0.93 per cent carbon 18-4-1 high speed steel were tested after treating from 2350 degrees Fahr. using (1) an atmosphere of 10 per cent CO or (2) an atmosphere of 2.5 per cent O_2.

(d) A micrographic comparison of 0.61, 0.70 and 0.76 per cent carbon 18-4-1 high speed steel specimens (⅜ x ½ x 1 inch) which were polished before hardening at 2350 degrees Fahr. Effect of a 10 per cent CO atmosphere and of a 2.5 per cent O_2 atmosphere at edges of sectioned originally polished surface unetched and etched.

(e) A comparison of the microhardness 0.0014 inch below the surface and 0.0154 inch below the surface of specimens (⅜ x ½ x 1 inch) of 0.61, 0.70 and 0.75 per cent carbon 18-4-1 high speed steels treated at 2350 degrees Fahr. in (1) an atmosphere of 10 per cent CO or (2) an atmosphere of 2.5 per cent O_2.

(f) The disposition of the carbon about 0.0005 inch below the treated surface to about 0.009 inch below the surface of a 0.80 per cent carbon molybdenum-vanadium high speed steel protected with borax in treating at 2200 degrees Fahr. in (1) an atmosphere of 11.7 per cent CO or (2) an atmosphere of 2.5 per cent O_2.

(g) A micrographic inspection of a molybdenum-tungsten high speed steel treated in an atmosphere of 34 to 35 per cent CO.

* * * * *

(A) Preliminary inspection of the effect of time (in the 2350 degrees Fahr. furnace, using an atmosphere of 10 per cent CO) on the carbon content of three successive surface cuts of 0.003 inch (0.006 inch on the diameter) using 0.70 per cent carbon 18-4-1 samples. Table V details the various data secured. Fig. 4 shows the curves obtained by plotting the carbon content of the various cuts against the time in the high temperature furnace. In the case of

the sample held in the 2350 degrees Fahr. furnace for 30 seconds, the carbon content of the first and second 0.003 inch cuts were lower than the carbon of the original sample. These points are indicated on the dotted lines as additional checks did not verify the result of

Fig. 4—Effect of Time (in the 2350 Degrees Fahr. Furnace Using an Atmosphere of 10 Per Cent CO) on the Disposition of the Absorbed Carbon in Three Successive 0.003-Inch Surface Cuts. Samples of 0.70 Per Cent Carbon 18-4-1 Steel (Heat "X") Were ⅞ Inch Diameter by 4 Inches Long. Preheat 1550 Degrees Fahr. 20 Minutes—2 Per Cent CO.

the first test. An additional sample of the 0.70 per cent carbon steel and a sample of the 0.60 per cent carbon steel were held for 30 seconds in the 2350 degrees Fahr. furnace with the results given in Table V. It is possible that there may be an incipient loss in carbon in the early period of heating in the 2350 degrees Fahr. furnace but the speed at which the surface reactions occur, as indicated by the carbon analyses of the various samples, would suggest difficulty in obtaining precise duplication of results. The curves show that this 18-4-1 steel has absorbed carbon in the first 0.003 inch of surface at an extremely rapid rate and at the end of 4 minutes the carbon was 0.85 per cent. The sample held 7 minutes in the furnace showed a lower carbon content on the first 0.003 inch of surface than the 4-minute sample. Carbon analyses of the three successive cuts on the 10-minute sample were all lower than the respective cuts of the 7-minute sample. The curves indicate a rapid absorption of carbon but after a certain period a decarburizing action occurs. The impli-

cation we may derive from this is that the hardened steel would have a modicum of soft surface even in the case of the 4-minute sample.

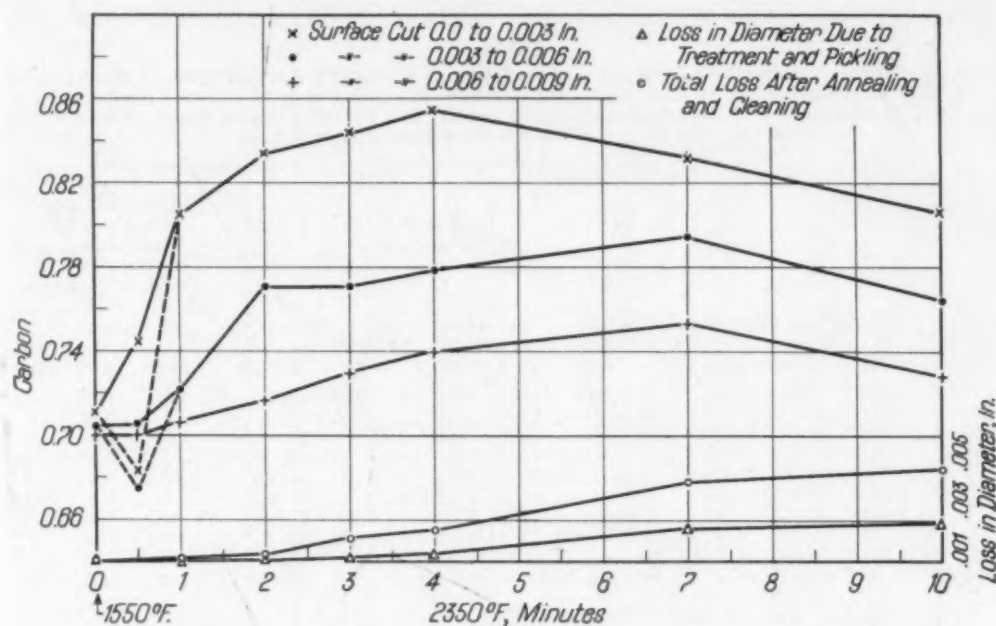In order to investigate the possible incipient loss of surface

Table V

Effect of Time (in the 2350 Degrees Fahr. Furnace Using an Atmosphere of 10 Per Cent CO) on the Disposition of the Absorbed Carbon in Successive Surface Cuts. Samples ⅞ Inch Diameter by 4 Inches Long, 0.70 Carbon 18-4-1 High Speed Steel (Heat "X"). Preheat 1550 Degrees Fahr.—20 Minutes, 2 Per Cent CO.

Sam-ple No.	Diameter of Samples Soft	Treat-ed	"An-nealed"	Time 2350° F.	As Quench-ed	Drawn 1050° F.	"An-neal-ed" 1425° F.	Run-out of Straightened Samples	Carbon Determinations Depth of Cuts in Respect to Surface of "An-nealed" Samples .000" to .003"	.003" to .006"	.006" to .009"
						Rockwell "C" Hardness					
X14	.8815"	.8815"	.8815"	None	19	18.5	18.2	Under .0005"	.716	.710	.710
X21	.8791"	.8791"	.8790"	30 seconds	53	50.0	31.0	Under .0005"	.684	.672	.710
X15	.8790"	.8790"	.8789"	1 minute	61.5	56.5	38.5	Under .0005"	.804	.730	.710
X16	.8795"	.8799"	.8793"	2 minutes	65.5	61.0	40.5	Under .0005"	.832	.770	.716
X17	.8790"	.8789"	.8780"	3 minutes	65.0	63.0	42.0	.0005"	.842	.770	.730
X18	.8670"	.8667"	.8655"	4 minutes	65.5	64.5	42.0	.0003"	.854	.778	.740
X19	.8801"	.8785"	.8762"	7 minutes	65.25	65.75	43.0	.0006"	.832	.794	.754
X20	.8790"	.8771"	.8745"	10 minutes	65.0	65.0	43.0	.0008"	.806	.764	.728
X11*	.8795"	.8796"	.8796"	30 seconds	52.0	....	....	.0007"	.744	.710	.702
Y11**	.8188"	.8187"	.8187"	30 seconds	49.0	....	....	.0005"	.644	.620	.606

*Sample X11 was 5" long.
**Sample Y11 was 0.60 carbon 18-4-1 steel 5" long.

Table VI

Experiment to Determine if an Incipient Decarburizing Action Occurs in the Early Heating of 18-4-1 Steel at 2350 Degrees Fahr. 0.71 Carbon 18-4-1 Steel of Heat "W" Used. Samples 5 Inches Long. 2350 Degrees Fahr. Atmosphere 10 Per Cent CO.

Sam-ple No.	Diameter of Samples Soft	Drawn	"An-nealed"	Treatment 1550 Deg. F.	2350 Deg. F.	Quench	Draw 1050 Deg. F.	Rockwell "C" Top End at ½ Radius	Mid-dle	1" from End
W2	1.6632"	1.6632"	1.6633"	40 Min.	None	Oil	1¼ Hours	24	19	..
W3	1.6421"	1.6428"	1.6427"	40 Min.	1¼ Min.	Oil	1¼ Hours	57	50.5	53
W4	1.7090"	1.7102"	1.7098"	40 Min.	2 Min.	Oil	1¼ Hours	61	55	..

Table VI—Continued

Sample No.	Area from Which Chips Were Obtained	Carbon Determinations Cuts in Respect to Surface of "Annealed" Samples .000" to .003"	.003" to .006"
W2	4" from unstamped end	.736	.708
W3	first 2" from unstamped end	.752	.718
W3	second 2" from unstamped end	.744	.720
W4	first 2" from unstamped end	.780	.738
W4	second 2" from unstamped end	.755	.712

carbon in the early period for heating at 2350 degrees Fahr. a further test was conducted using larger diameter samples of heat "W". Samples approximately 1⅛ inch diameter by 5 inches long were treated as follows: Sample W2 heated 1550 degrees Fahr. quenched in oil; sample W3 preheated 1550 and held in the 2350 degrees Fahr.

Fig. 5—Probability Curve Showing the Disposition of the Carbon with Respect to the Surface of Hardened, Drawn and Pickled 0.60 Per Cent Carbon 18-4-1 High Speed Steel. Samples ⅞ Inch Diameter by 5 Inches Long. Preheat (1550 Degrees Fahr.) Atmosphere 2 Per Cent CO; High Heat (2350 Degrees Fahr.) Atmosphere 10 Per Cent CO.

furnace for 1¼ minutes and sample W4 preheated and held in the 2350 degrees Fahr. furnace for 2 minutes. Table VI shows the results of the carbon determinations made on successive surface cuts. On sample W2, given the 1550 degrees Fahr. heat only, chips were obtained 0.000 to 0.003 inch and 0.003 to 0.006 inch in respect to the "annealed" and cleaned surface. On samples W3 and W4 chips 0.000 to 0.003 inch and 0.003 to 0.006 inch were obtained from each on the first 2 inches from the unstamped end. Chips to the same depth of cuts were secured on the second 2-inch length from the unstamped end. The carbon analyses show no loss in carbon compared to the original carbon of samples. The analyses do show that the carbon has been absorbed at a faster rate on the first 2-inch length of sample as compared to the second 2-inch length. The top-

most end of sample would heat faster than the middle portion of sample and a faster rate of carbon absorption might be expected.

(B)   Disposition of the carbon in the zone approximately 0.001 inch below the surface to about 0.012 inch below the surface of treated round samples of 0.60, 0.70 and 0.93 per cent carbon 18-4-1 high speed steel.   Table VII gives the various data such as size change due to treatment and to lead "annealing" at 1425 degrees Fahr., the hardness, distortional changes and carbon analyses of the
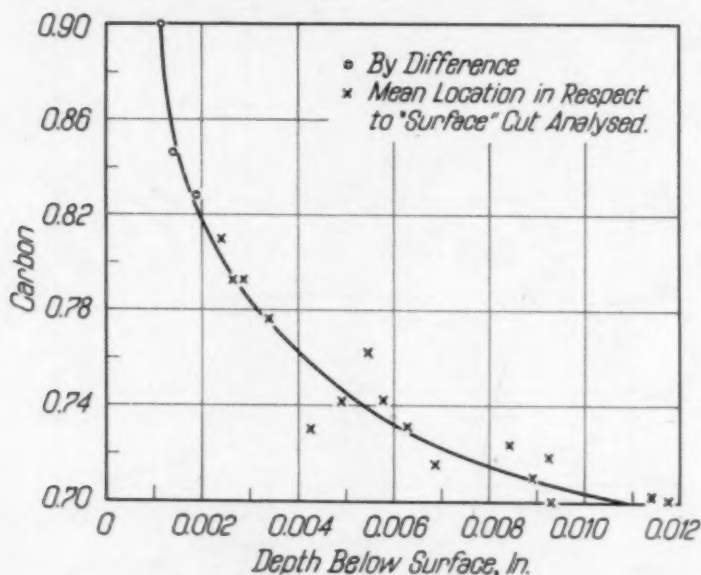
Fig. 6—Probability Curve Showing the Disposition of the Carbon with Respect to the Surface of Hardened, Drawn and Pickled 0.70 Per Cent Carbon 18-4-1 High Speed Steel. Samples 1⅛ Inch Diameter by 5 Inches Long. Preheat (1550 Degrees Fahr.) Atmosphere 2 Per Cent CO; High Heat (2350 Degrees Fahr.) Atmosphere 10 Per Cent CO.

various surface cuts.   All samples were treated at 2350 degrees Fahr. for 4 minutes total time, using an atmosphere of either 10 per cent CO or 2.5 per cent O_2.   The carbon analyses refer to the cuts made in reference to the surface of the 1425 degrees Fahr. "annealed" and cleaned samples.

Figs. 5, 6, 7, and 8 show respectively the plotted carbon values against the depth below the surface of 0.60, and 0.70 per cent carbon 18-4-1 steels treated at 2350 degrees Fahr. in an atmosphere of 10 per cent CO and the 0.60 and 0.70 per cent carbon steels treated at 2350 degrees Fahr. in an atmosphere of 2.5 per cent O_2.   The locations of all points have been corrected as regards position with respect to the treated surface.   The first three points on each curve have been obtained by difference and located at the mean of the cal-
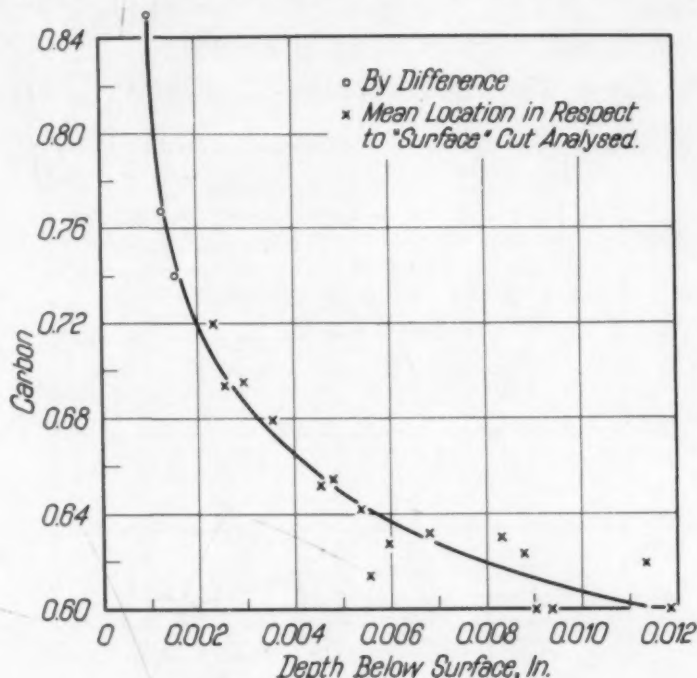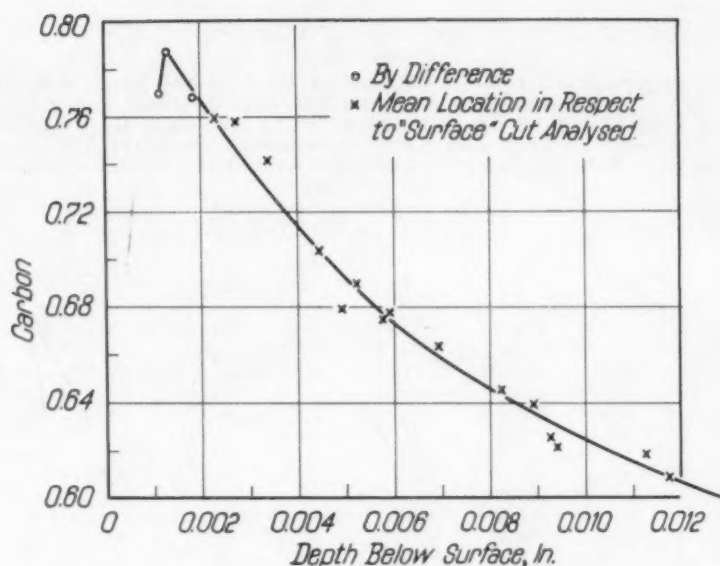
Fig. 7—Probability Curve Showing the Disposition of the Carbon with Respect to the Surface of Hardened, Drawn and Pickled 0.60 Per Cent Carbon 18-4-1 High Speed Steel. Samples ¹³⁄₁₆ Inch Diameter by 5 Inches Long. Preheat (1550 Degrees Fahr.) Atmosphere 2 Per Cent CO; High Heat (2350 Degrees Fahr.) Atmosphere 2.5 Per Cent O_2.
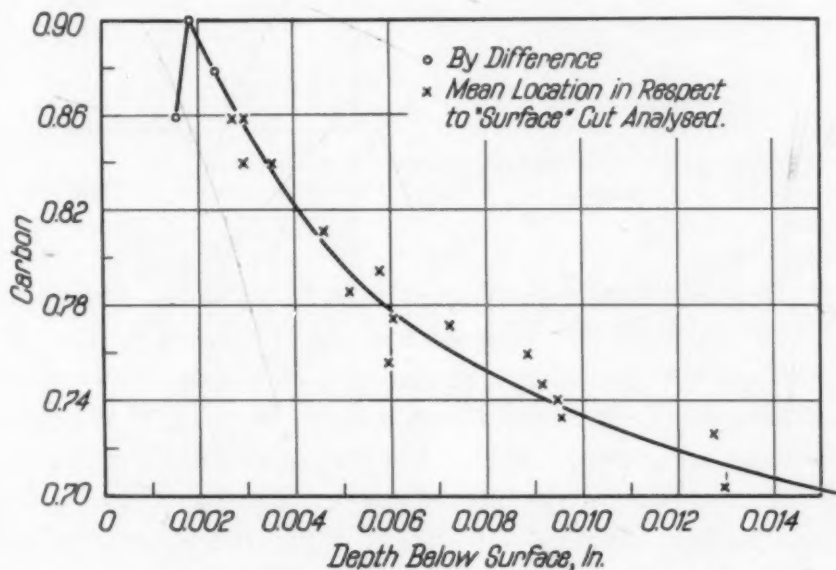
Fig. 8—Probability Curve Showing the Disposition of the Carbon with Respect to the Surface of Hardened, Drawn and Pickled 0.70 Per Cent Carbon 18-4-1 High Speed Steel. Samples ⅞ Inch Diameter by 5 Inches Long. Preheat (1550 Degrees Fahr.) Atmosphere 2 Per Cent CO; High Heat (2350 Degrees Fahr.) Atmosphere 2.5 Per Cent O_2.

culated zone. All other points on the curves are positioned by using the mean location in respect to the various cuts made but corrected to the treated surface as mentioned above. Inspection of the curves show we have no analytical data of the zone from the surface to

Table VII

Various Data on Approximately ⅞ Inch Diameter by 5 Inches Long, 0.60, 0.70 and 0.93 Carbon 18-4-1 High Speed Steels Hardened from 2350 Degrees Fahr. Using (1) an Atmosphere of 10 Per Cent CO or (2) an Atmosphere of 2.5 Per Cent O₂. Constant Preheat (1550 Degrees Fahr.—2 Per Cent CO) Time 20 Minutes. Constant 2350 Degrees Fahr. Time 4 Minutes. Data on Samples Quenched from 1550 Degrees Fahr. Furnace Also Included

| Sample No.	Carbon of Heat	Diameter of Samples			Atmosphere 2350 Deg. F.	Rockwell "C" After 1050 Deg. F. 1½ Hours	Distortion Due to Treatment	Run-out of Samples After Straightening
		Soft	Treated	"Annealed"				
Y-4	.60	.8185"	.8180"	.8165"	10% CO	63.8	.0110"	.0015"—.0010"—.0005"
Y-6	.60	.8143"	.8130"	.8113"	2.5% O₂	64.0	.0130"	.0020"—.0000"—.0000"
X-14	.70	.8785"	.8782"	.8764"	10% CO	65.0	.0160"	.0020"—.00125"—.0015"*
X-8	.70	.8777"	.8758"	.8735"	2.5% O₂	65.5	.0140"	.0015"—.0000"—.0000"
Z-3	.93	.8793"	.8770"	.8756"	10% CO	66.0	.0075"	.0015"—.0015"—.0015"*
Z-6	.93	.8790"	.8766"	.8744"	2.5% O₂	66.0	.0045"	.00125"—.0005"—.0005"
Y-3	.60	.8119"	.8114"	.8099"	10% CO	64.0	.0090"	.0000"—.0010"—.0005"
Y-9	.60	.8192"	.8177"	.8158"	2.5% O₂	64.5	.0150"	.00125"—.0010"—.0010"*
X-4	.70	.8802"	.8800"	.8785"	10% CO	65.0	.0135"	.0020"—.0001"—.0010"
X-7	.70	.8806"	.8796"	.8774"	2.5% O₂	65.75	.0210"	.0007"—.0005"—.0005"
Z-2	.93	.8792"	.8763"	.8750"	10% CO	66.0	.0040"	.0020"—.0010"—.0010"*
Z-5	.93	.8801"	.8766"	.8752"	2.5% O₂	66.0	.0060"	.0007"—.0005"—.0000"
Y-2	.60	.8180"	.8172"	.8155"	10% CO	63.8	.0120"	.0005"—.0007"—.0011"
Y-8	.60	.8177"	.8153"	.8138"	2.5% O₂	64.25	.0110"	.0020"—.0015"—.0010"*
X-1	.70	.8770"	.8767"	.8752"	10% CO	65.0	.0145"	.0015"—.0005"—.0020"
X-6	.70	.8703"	.8688"	.8670"	2.5% O₂	65.75	.0200"	.00125"—.0005"—.0020"
Z-1	.93	.8662"	.8632"	.8613"	10% CO	66.0	.0035"	.0005"—.0005"—.0005"
Y-1	.60	.8187"	.8182"	.8163"	10% CO	64.0	.0130"	.0025"—.0010"—.0025"*
Y-7	.60	.8178"	.8163"	.8143"	2.5% O₂	64.0	.0190"	.0030"—.0005"—.0003"
X-2	.70	.8745"	.8740"	.8723"	10% CO	65.0	.0170"	.0020"—.0015"—.0015"*
X-9	.70	.8777"	.8764"	.8744"	2.5% O₂	66.0	.0130"	.0004"—.0000"—.0010"
Y-12	.60	.8136"	.8135"	.8135"	1550° F.	19.0	.0003"	.0000"—.0000"—.0000"
Y-13	.60	.8115"	.8115"	.8114"	1550° F.	19.0	.0003"	.0000"—.0003"—.0000"
P-12	.70	.8806"	.8806"	.8805"	1550° F	20.5	.0003"	.0000"—.0005"—.0000"
P-13	.70	.8784"	.8784"	.8783"	1550° F	20.5	.0004"	.0000"—.0006"—.0000"

*Weave in samples did not permit improvement in straightness.

about 0.001 inch below the surface. Perhaps the most significant point in connection with these curves is that the slopes of the curves for the 0.60 and 0.70 per cent carbon steels treated in the 10 per cent CO atmosphere are about the same. The slopes of curves of the 0.60 and 0.70 per cent carbon steel treated in the 2.5 per cent O₂ atmosphere are also comparable. The slope of the curve of the 10 per cent CO atmosphere samples is steeper than the slope of the 2.5 per cent atmosphere samples. This might imply that some or more surface heating (1)[1] occurs on the 2.5 per cent O₂ samples as compared to the 10 per cent CO atmosphere samples. A more rapid diffusion of the carbon and a greater scaling would be a natural consequence.

All of the 0.93 per cent carbon 18-4-1 steel samples showed about 0.03 per cent "pick-up" of carbon in the first 0.003 inch of "surface" whether treated in the 10 per cent CO atmosphere or in the 2.5 per cent O₂ atmosphere. There is little or no diffusion of carbon beyond a depth of 0.003 inch below the treated surfaces.

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[1]The figures appearing in parentheses refer to the bibliography appended to this paper.

Table VII—Continued

Sample No.	Carbon of Heat	Atmosphere 2350 Deg. Fahr.	Surface Removed by Polishing or Grinding	Comment on Ground Surfaces	.000"/.003"	.003"/.006"	.006"/.009"	.009"/.012"
Y-4	.60	10% CO	None	....	.720	.642	.630	.62
Y-6	.60	2.5% O₂	None	....	.760	.686	.646	.63
X-14	.70	10% CO	None	....	.810	.762	.724	.702
Z-8	.70	2.5% O₂	None	....	.858	.794	.758	.726
Z-3	.93	10% CO	None	....	.955	.931	.927	.93
Z-6	.93	2.5% O₂	None	....	.955	.940	.941	.925
Y-12	.60	1550 deg. F. only 20 Min.	None	....	.630	.606	.604	....
Y-13	.60		None	....	.630			....
X-12	.60		None	....	.71	.704	.704	....
X-13	.70		None	....	.73			....

Sample No.	Carbon of Heat	Atmosphere 2350 Deg. Fahr.	Surface Removed by Polishing or Grinding	Comment on Ground Surfaces	.0005"/.003"	.003"/.005"	.005"/.007"	.007"/.009"
Y-3	.60	10% CO	.0005" by polishing	....	.694	.654	.63	.624
Y-9	.60	2.5% O₂	.0005" by polishing	....	.758	.68	.664	.64
X-4	.70	10% CO	.0005" by polishing	....	.792	.742	.716	.71
X-7	.70	2.5% O₂	.0005" by polishing	....	.858	.784	.770	.748
Z-2	.93	10% CO	.0005" by polishing	....	.956	.952	.936	.926
Z-5	.93	2.5% O₂	.0005" by polishing	....	.960	.948	.930	.930

Sample No.	Carbon of Heat	Atmosphere 2350 Deg. Fahr.	Surface Removed by Polishing or Grinding	Comment on Ground Surfaces	.001"/.003"	.003"/.007"	.007"/.010"	.010"/.012"
Y-2	.60	10% CO	.001" by grinding	95% Bright	.696	.628	.600	.600
Y-8	.60	2.5% O₂	.001" by grinding	80% Bright	.746	.676	.626	.61
X-1	.70	10% CO	.001" by grinding	90% Bright	.792	.742	.718	.696
X-6	.70	2.5% O₂	.001" by grinding	85% Bright	.838	.755	.740	.704
Z-1	.93	10% CO	.001" by grinding	100% Bright	.954	.932	.928	.932

Sample No.	Carbon of Heat	Atmosphere 2350 Deg. Fahr.	Surface Removed by Polishing or Grinding	Comment on Ground Surfaces	.002"/.003"	.003"/.004"	.004"/.006"	.006"/.011"
Y-1	.60	10% CO	.002" by grinding	100% Bright	.680	.653	.614	.60
Y-7	.60	2.5% O₂	.002" by grinding	100% Bright	.742	.703	.678	.622
X-2	.70	10% CO	.002" by grinding	100% Bright	.776	.73	.73	.696
X-9	.70	2.5% O₂	.002" by grinding	100% Bright	.838	.81	.774	.732

(The four carbon-determination sub-columns in each group are headed "Carbon Determinations Cuts in Reference to the Surface of the "Annealed" Samples.")

In Table VIII some af the various data of Table VII are reduced to averages in regard to samples of the same steel given similar processing. On the 0.60 and 0.70 per cent carbon 18-4-1 steels the use of a 10 per cent CO atmosphere compared with the 2.5 per cent O₂ atmosphere shows (1) a somewhat lower hardness, (2) somewhat less distortion and (3) a loss in diameter of the order of 1 to 3. The averages of the 0.93 per cent carbon 18-4-1 steel show the hardness,

Table VIII

Various Data of Table V Reduced to Averages of the Respective Samples of the 0.60, 0.70 and 0.93 Carbon 18-4-1 High Speed Steels Which Were Given Similar Processing. The 0.60 Carbon Samples Were ¹⅜ Inch Diameter; the 0.70 and 0.93 Carbon Samples Were ⅞ Inch Diameter

	0.60 Carbon 18-4-1 2350 Deg. F. Atmosphere		0.70 Carbon 18-4-1 2350 Deg. F. Atmosphere		0.93 Carbon 18-4-1 2350 Deg. F. Atmosphere	
	10% CO	2.5% O_2	10% CO	2.5% O_2	10% CO	2.5% O_2
No. Samples Averaged	4	4	4	4	3	2
Distortion	.0111"	.0150"	.0155"	.0170"	.0050"	.0051"
Loss in size due to hardening	.00056"	.0017"	.00042"	.0014"	.0027"	.0029"
Loss of treated diameter due to "annealing and cleaning	.00165"	.0018"	.0016"	.0021"	.0015"	.0013"
Rockwell "C"	63.9	64.2	65.0	65.7	66.0	66.0

Table IX

Influence of a 10 Per Cent CO Atmosphere Compared with an Atmosphere of 2.5 Per Cent O_2 on the Disposition of the Absorbed Carbon in the Surface Zone After Hardening .60, .70 and .93 Carbon 18-4-1 High Speed Steels at 2350 Degrees Fahr., the Samples Having Been Coated with Borax. Samples Preheated 1550 Degrees Fahr.—10 Minutes, Removed and the Borax Applied. Samples Returned to Preheat for an Additional 12 Minutes

Sample No.	Carbon of Heat	Diameter of Samples			2350 Deg. F. Time— 4 Min. Atmosphere	Rockwell "C" After 1050° F. 1½ Hours	Distortion Due to Treatment	Run-out of Samples After Straightening
		Soft	Treated	Annealed				
Y5	.60	.8018"	.8005"	.7993"	10% CO	64.0	.0140"	.00175"—.0005"—.0010"
Y10	.60	.8141"	.8114"	.8088"	2.5% O_2	65.25	.0120"	.0000" —.0010"—.0020"
X5	.70	.8783"	.8770"	.8754"	10% CO	65.0	.0105"	.0000" —.0010"—.0010"
X10	.70	.8790"	.8756"	.8744"	2.5% O_2	66.5	.0120"	.0020" —.0020"—.0010"*
Z4	.93	.8793"	.8755"	.8750"	10% CO	66.0	.0050"	.0005" —.0020"—.0000"
Z7	.93	.8776"	.8725"	.8720"	2.5%O_2	66.0	.0075"	.0010" —.0010"—.0000"

Table IX—(Continued)

Sample No.	Carbon of Heat	2350 Deg. F. Atmosphere	Carbon Determinations Surface Cuts in Reference to "Annealed" Surface		
			.000 to .003"	.003 to .006"	.006 to .009"
Y5	.60	10% CO	.734	.650	.632
Y10	.60	2.5% O_2	.788	.690	.672
X5	.70	10% CO	.818	.742	.740
X10	.70	2.5% O_2	.866	.804	.762
Z4	.93	10% CO	.964	.934	.922
Z7	.93	2.5% O_2	.948	.928	.928

*Weave in sample did not permit improvement in straightness.

distortion and loss in size due to hardening to be about the same in either atmosphere. The average distortion of the 0.93 per cent carbon samples is about one third that of the 0.70 per cent carbon samples. The implication of this marked difference in distortion between the 0.70 and 0.93 per cent carbon steels together with the fact that the 0.93 per cent carbon steel absorbs relatively little carbon in treatment, may be that the carbon absorption contributes to distortion. We could hardly anticipate a uniform absorption of carbon

about the periphery of a round sample placed vertically in the 2350 degree Fahr. furnace.

(C) Table IX gives the various data obtained on 0.60, 0.70 and 0.93 per cent carbon 18-4-1 steels which were borax coated before placing in the 2350 degree Fahr. furnace. Samples were removed from the preheat furnace after reaching about 1200 degrees Fahr.
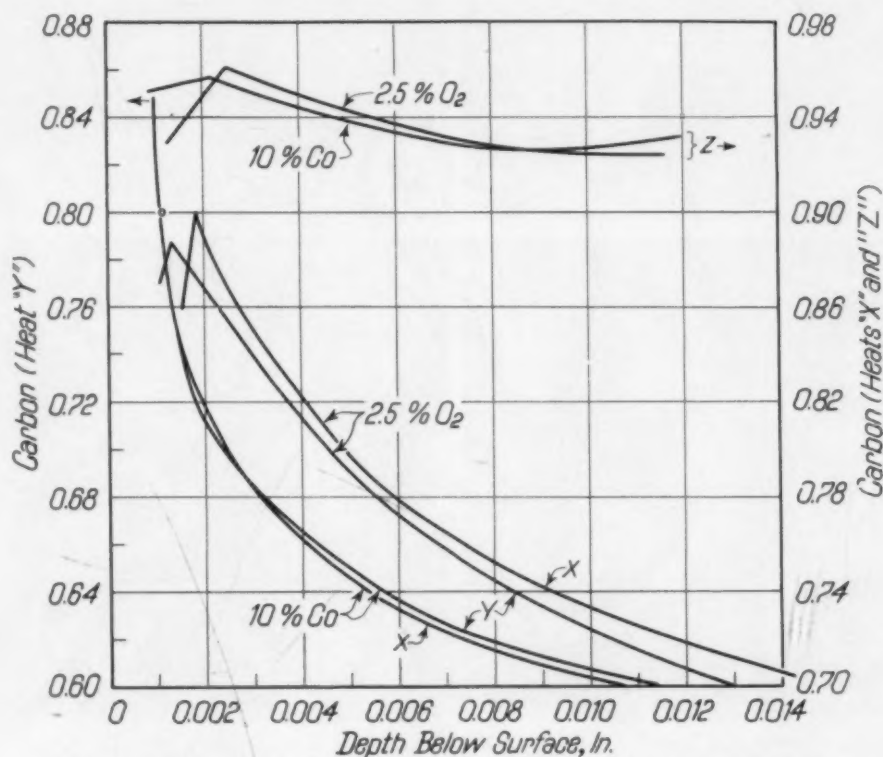

Fig. 9—Composite of Curves (From Data of Table VII) Showing the Probable Disposition of the Carbon from Approximately 0.001 Inch Below the Treated Surface to About 0.012 Inch Below the Surface of 0.60, 0.70 and 0.93 Per Cent Carbon 18-4-1 Steels Hardened from 2350 Degrees Fahr. in (1) an Atmosphere of 10 Per Cent CO and (2) an Atmosphere of 2.5 Per Cent O_2.

(or after 10 minutes in the furnace) coated with borax and then returned to the preheat for an additional 12 minutes. The use of borax made no marked change in the carbon contents of the various cuts when compared atmosphere for atmosphere with the respective steels not borax coated.

(D) Figs. 10 to 21 inclusive are a photomicrographic comparison of sections of treated surfaces of approximately 0.60, 0.70 and 0.75 per cent carbon 18-4-1 steels hardened from 2350 degrees Fahr. using (1) an atmosphere of 10 per cent CO or (2) an atmosphere of 2.5 per cent O_2. Specimens approximately $\frac{3}{8}$ x $\frac{1}{2}$ x 1 inch were polished on one $\frac{1}{2}$ x 1 inch side before hardening. Specimens pre-
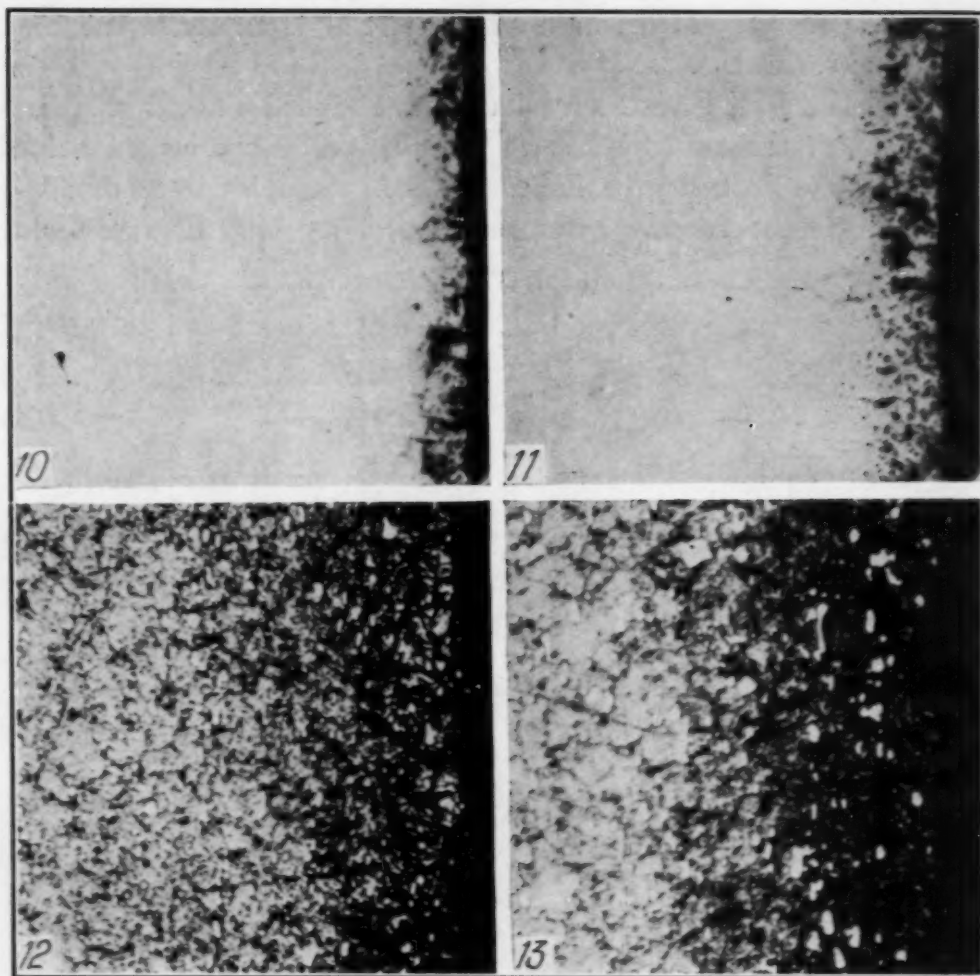
Fig. 10—Sample C1. 2350 Degrees Fahr. Atmosphere. 10 Per Cent CO. Unetched.
× 500.
Fig. 11—Sample C4. 2350 Degrees Fahr. Atmosphere. 2.5 Per Cent O_2. Unetched.
× 500.
Fig. 12—Sample C1. 2350 Degrees Fahr. Atmosphere. 10 Per Cent CO. Etched
4 Per Cent Nital 45 Seconds. × 500.
Fig. 13—Sample C4. 2350 Degrees Fahr. Atmosphere. 2.5 Per Cent O_2. Etched
4 Per Cent Nital 50 Seconds. × 500.
Figs. 10 to 13 Inclusive—Sections at Surface of a 0.61 Per Cent Carbon 18-4-1
Steel Hardened from 2350 Degrees Fahr. in Either a 10 Per Cent CO or a 2.5 Per Cent
O_2 Atmosphere. Specimens Drawn 1050 Degrees Fahr.—1.5 Hours Redrawn 1080 De-
grees Fahr.—1.5 Hours.

pared from the various heats of 18-4-1 steel given below were used
either for micrographic comparison or for the microhardness tests
under section (E).

| Heat Symbol	C	Mn	P	S	Si	Cr	W	V
"C"	0.61	0.26	0.026	.....	0.26	4.00	18.17	1.11
"M"	0.61	0.26	0.017	0.011	0.08	3.99	17.53	1.09
"N"	0.70	0.20	0.019	tr	0.28	3.90	18.04	1.02
"Q"	0.70	0.19	0.019	tr	0.35	4.00	17.76	0.99
"S"	0.76	0.21	0.015	tr	0.36	3.93	17.96	1.04
"T"	0.75	0.26	0.018	tr	0.39	3.96	17.64	1.03

Fig. 14—Sample N1. 2350 Degrees Fahr. Atmosphere. 10 Per Cent CO. Unetched. × 500.

Fig. 15—Sample N4. 2350 Degrees Fahr. Atmosphere. 2.5 Per Cent O_2. Unetched. × 500.

Fig. 16—Sample N1. 2350 Degrees Fahr. Atmosphere. 10 Per Cent CO. Etched 4 Per Cent Nital 1 Minute, 15 Seconds. × 500.

Fig. 17—Sample N4. 2350 Degrees Fahr. Atmosphere. 2.5 Per Cent O_2. Etched 4 Per Cent Nital 45 Seconds. × 500.

Figs. 14 to 17 Inclusive—Sections at Surface of a 0.70 Per Cent Carbon 18-4-1 Steel Hardened from 2350 Degrees Fahr. in Either a 10 Per Cent or a 2.5 Per Cent O_2 Atmosphere. Specimens Drawn 1050 Degrees Fahr.—1.5 Hours, Redrawn 1080 Degrees Fahr.—1.5 Hours.

All specimens were held in the 2350 degrees Fahr. furnace for a total time of 2 minutes 15 seconds. All specimens were quenched in oil and then degreased with benzol. The specimens after drawing at 1050 degrees Fahr.—1½ hours and redrawing at 1080 degrees Fahr.—1½ hours were cold pickled. The surface which had been polished before hardening was cleaned to a metallic appearance by means of an abrasive eraser. Samples were then nicked in two and the section at the surface of the original polished and hardened sur-
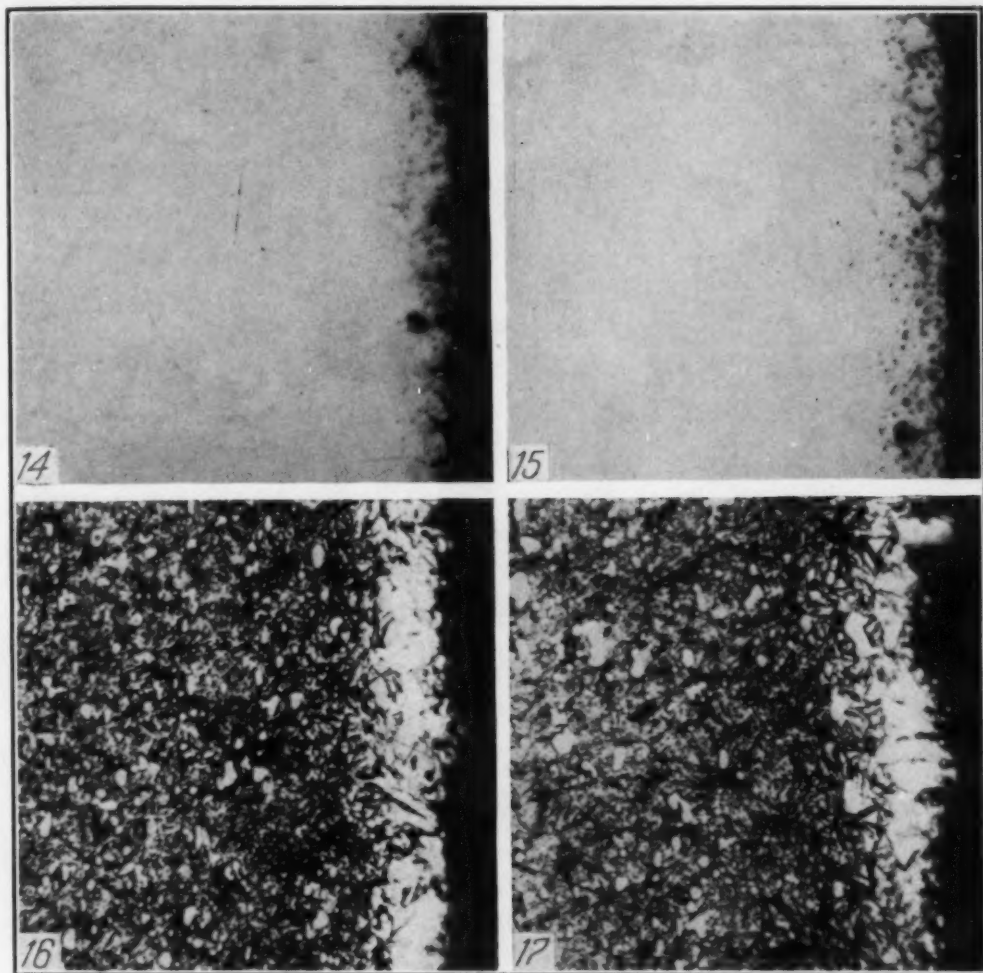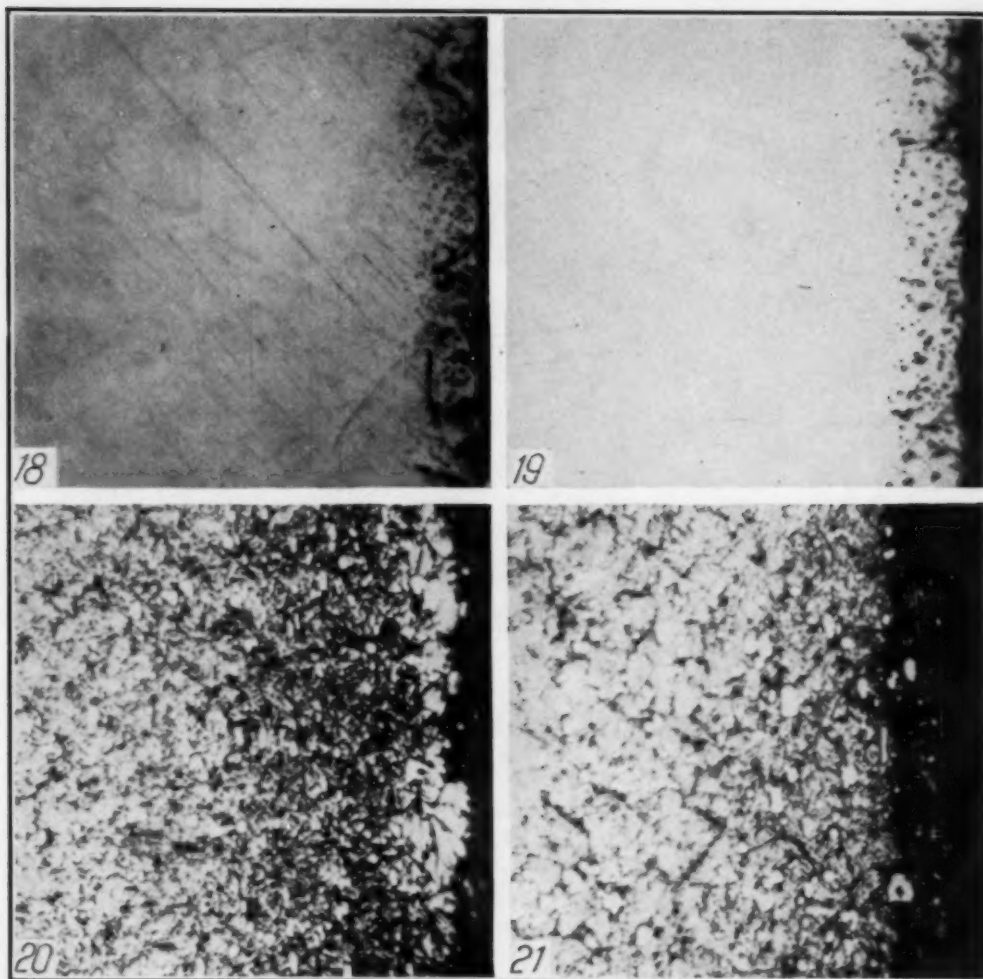
Fig. 18—Sample S1. 2350 Degrees Fahr. Atmosphere. 10 Per Cent CO. Unetched. × 500.
Fig. 19—Sample S4. 2350 Degrees Fahr. Atmosphere. 2.5 Per Cent O₂. Unetched. × 500.
Fig. 20—Sample S1. 2350 Degrees Fahr. Atmosphere. 10 Per Cent CO. Etched 4 Per Cent Nital 45 Seconds. × 500.
Fig. 21—Sample S4. 2350 Degrees Fahr. Atmosphere. 2.5 Per Cent O₂. Etched 4 Per Cent Nital 45 Seconds. × 500.
Figs. 18 to 21 Inclusive—Sections at Surface of a 0.76 Per Cent Carbon 18-4-1 Steel Hardened from 2350 Degrees Fahr. in Either a 10 Per Cent CO or a 2.5 Per Cent O₂ Atmosphere. Specimens Drawn 1050 Degrees Fahr.—1.5 Hours, Redrawn 1080 Degrees Fahr.—1.5 Hours.

face micrographed unetched and etched at × 500.

All photomicrographs of the unetched specimens reveal a more or less pitted surface to a depth of about 0.0005 inch.

The 0.61 per cent carbon specimens after etching show a dark outer zone. This zone is deeper and more intense in the specimen treated from the 2.5 per cent O₂ atmosphere.

The 0.70 per cent carbon 18-4-1 specimens after etching show massive carbide at the surface to a depth of approximately 0.0005

inch. The 2.5 per cent O_2 atmosphere specimen reveals this zone as less continuous as well as less intense as compared to the 10 per cent CO atmosphere specimen.

The 0.76 per cent 18-4-1 specimen hardened from the 10 per cent CO atmosphere shows a small amount of massive carbide at the surface. The 2.5 per cent O_2 atmosphere specimen shows no massive carbide at the surface.

The 0.61 per cent carbon specimens in the short time employed at 2350 degrees Fahr. probably did not absorb enough carbon at the surface to form massive carbide. That the higher carbon exterior of the 0.61 per cent carbon specimens etch darker than the interior of sample is compatible with the normal resistance to darkening of

Table X

Comparison of the Rockwell and Microhardness (Knoop) Numbers Taken 0.0014 and 0.0154 inch Below the Hardened and Drawn Surfaces of 0.61, 0.70 and 0.75 Per Cent Carbon 18-4-1 High Speed Steels. Specimens ½ by ⅜ by 1 Inch Were Drawn 1050 Degrees Fahr. —1½ Hours; Redrawn 1080 Degrees Fahr.—1½ Hours

Specimen No.	Carbon	Atmosphere 2350 Deg. F.	Depth Below the Treated Surface ± .0002"	Rockwell "C"	Microhardness Numbers (Loads in Grams)		
					2000	1000	500
M-1	.61	10% CO	.0014"	63.3	781	806	795
			.0154"	63.1	723	726	744
M-4	.61	2.5% O_2	.0014"	63.3	790	832	861
			.0154"	63.1	736	734	753
Q-1	.70	10% CO	.0014"	65.1	833	825	870
			.0154"	65.0	764	799	812
Q-4	.70	2.5% O_2	.0014"	65.1	831	845	907
			.0154"	65.0	803	818	808
T-1	.75	10% CO	.0014"	65.2	835	849	861
			.0154"	65.2	801	823	829
T-4	.75	2.5% O_2	.0014"	65.1	809	799	855
			.0154"	65.3	803	793	822

the matrix of a 0.61 per cent carbon 18-4-1 steel drawn as were these specimens.

It is probable the 0.76 per cent carbon specimen treated in the 2.5 per cent O_2 atmosphere reached about 0.95 per cent carbon at the surface more rapidly than the specimen treated in the 10 per cent CO atmosphere and the decarburizing action then proceeded with the result that the carbon content at the surface was too low to form massive carbide.

(E) Microhardness (Knoop) numbers were secured 0.0014 and 0.0154 inch below the hardened and drawn surfaces of 0.61, 0.70 and 0.75 per cent carbon 18-4-1 high speed steel specimens. The specimens (½ x ⅜ x 1 inch) were hardened from 2350 degrees Fahr. in (1) an atmosphere of 10 per cent CO or (2) an atmosphere of 2.5 per cent O_2. The specimens were prepared for

hardening by polishing one ½ x 1 inch side; the opposite side was ground only. Specimens were hardened (a ⅜-inch side resting on the hearth) and double drawn. After pickling 0.0010 inch was ground from the side which had been polished before hardening; 0.0150 inch was ground from the opposite side. Microhardness tests at 500, 1000 and 2000 gram loads were determined on each side after polishing. The polishing removed an additional 0.0004 inch plus or minus 0.0002 inch. The results of the microhardness tests are tabulated in Table X.

On the 0.61, 0.70 and 0.75 per cent carbon specimens the microhardness numbers taken 0.0014 inch below the treated surfaces are definitely higher than the numbers obtained 0.0154 inch below the treated surface on the same specimen. The hardness numbers for the 0.61 and 0.70 carbon steels are higher 0.0014 inch below the surface when the respective steels hardened from the 2.5 per cent O_2 atmosphere are compared with the 10 per cent CO atmosphere. The higher hardness obtained 0.0014 inch below the treated surfaces as compared with the hardness 0.0154 inch below the treated surface on the same specimen would appear a natural consequence of the carbon absorption which occurs in the 2350 degrees Fahr. furnace.

The microhardness test using the Knopp indenter is extremely sensitive. The size and distribution of the carbides, which Knoop, Peters and Emerson (2) have found to show a microhardness of about 1400, influence the values somewhat. It is believed, however, the values listed in Table X give a relatively true comparison of the difference in hardness between the two levels tested on the same specimen.

(F)   Disposition of the carbon from about 0.0005 inch below the surface to about 0.009 inch below the surface of treated 0.80 per cent carbon-molybdenum-vanadium high speed steel (heat "V") protected with borax in hardening from 2200 degrees Fahr. in (1) an atmosphere of 11.7 per cent CO, or (2) an atmosphere of 2.5 per cent O_2.

The borax was applied as indicated in Table XI. After hardening and quenching in oil all samples (except Nos. 2 and 3) were "annealed" in lead at 1400 degrees Fahr. for ten minutes, cooled in silocel and then sandblast cleaned. Samples Nos. 2 and 3 were quenched in distilled water from the 1450 degrees Fahr. furnace and were not "annealed" or sandblast cleaned. These two samples were cleaned in benzol only before machining. The residue result-

Table XI

Comparison of ¾ Inch Diameter by 5 Inches Long 0.80 Carbon-Molybdenum-Vanadium High Speed Steel (Heat "V") as Regards the Disposition of the Carbon from the Surface to About 0.009 Inch Below the Surface After Hardening from 2200 Degrees Fahr. in (1) an Atmosphere of 11.7 Per Cent CO or (2) an Atmosphere of 2.5 Per Cent O_2

Sample No.	Diameter of Samples			Treatment	Time 2200° F.	2200° F. Atmosphere	Quench	Rockwell "C" After 1025° F. 2 Hours
	Soft	Treated	"An-nealed"	Time 1450° F. 2% CO				
2*	.7570"	.7570"	.....	10 minutes	None	......	Distilled Water	....
3*	.7590"	.75925"	.....	{10 min. Borax {22 min. Total	None	......	Distilled Water	....
9	.75925"	.7565"	.7562"	5 min. Borax	5 min.	11.7% CO	Oil	65.0
7	.75825"	.7560"	.7555"	{10 min. Borax {22 min. Total	4 min.	11.7% CO	Oil	65.0
8	.7575"	.7555"	.7543"	{10 min. Borax {22 min. Total	4 min.	11.7% CO	Oil	64.8
1	.7585"	.7550"	.7536"	5 min. Borax	5 min.	2.5% O_2	Oil	65.1
4	.7585"	.75475"	.7535"	{10 min. Borax {22 min. Total	4 min.	2.5% O_2	Oil	64.9
6	.75675"	.7530"	.7520"	{10 min. Borax {22 min. Total	4 min.	2.5% O_2	Oil	65.1

Table XI—(Continued)

Sample No.	2200° F. Atmosphere	Carbon Determinations Surface Cuts in Respect to Surfaces of "Annealed" Samples					
		.000 to .002"	.000 to .003"	.002 to .004"	.003 to .006"	.004 to .006"	.006 to .009"
2	....	.836	....	....	....	....	....
3	....	.836	....	....	....	....	....
9	11.7% CO	....	.882	....	.836	....	.826
7	11.7% CO	....	.885	....	.826	....	.820
8	11.7% CO	.902	....	.874	....	.836	....
1	2.5% O_2	....	.934	....	.854	....	.834
4	2.5% O_2	....	.938	....	.850	....	.822
6	2.5% O_2	.950	....	.874	....	.836	....

*No. 2 and No. 3 samples were not "annealed" or sandblast cleaned; cleaning was only in benzol.

ing from quenching No. 3 sample in distilled water was treated with HCl. Examination showed no indication of any soot. The residue showed the presence of iron indicating the dissolving action of borax at the 1450 degrees Fahr. temperature. No residue was obtained from quenching the No. 2 sample. After straightening and cleaning in benzol various surface cuts were made and the chips analyzed for carbon with the results tabulated in Table XI.

Carbon absorption occurs in either atmosphere and as in the case of the 18-4-1 steel the absorption is greater in the 2.5 per cent O_2 atmosphere than in the high CO atmosphere. The difference as regards "surface" carbon content between the samples of the molybdenum steel treated in the 11.7 per cent CO atmosphere and in the 2.5 per cent O_2 atmosphere is not as great as the difference between

the 0.60 per cent carbon 18-4-1 steel samples treated at 2350 degrees Fahr. in the 10 per cent CO atmosphere and in the 2.5 per cent O_2 atmosphere; this is also true in the case of the 0.70 per cent carbon steel.
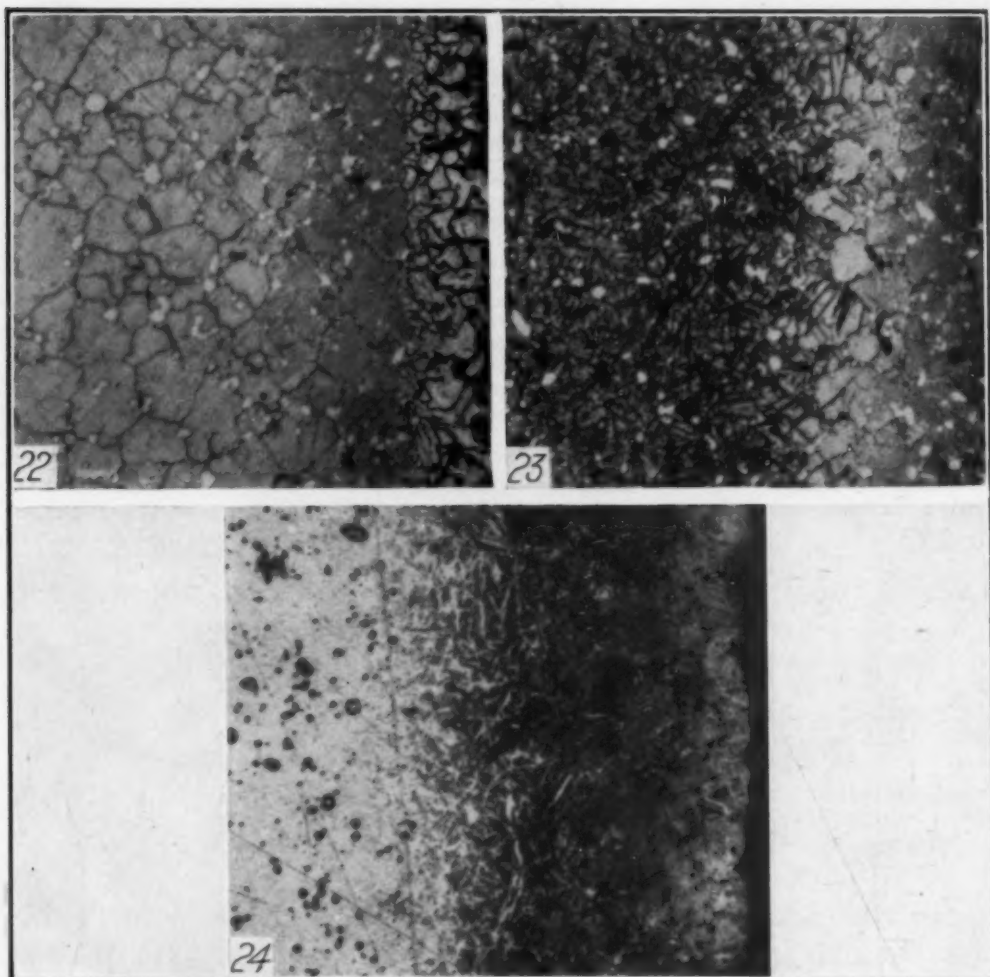

Fig. 22—Sample No. 2. As Quenched. Etched 4 Per Cent Nital. × 500.
Fig. 23—Sample No. 4. Drawn 1025 Degrees Fahr.—1 Hour. Etched 4 Per Cent Nital. × 500.
Fig. 24—Sample No. 4. Drawn 1025 Degrees Fahr.—1 Hour. Etched in Boiling Sodium Picrate. × 500.
Figs. 22 to 23 Inclusive—Sections at the Hardened Surface of a 0.75 Per Cent Molybdenum-Tungsten High Speed Steel Treated at 2240 Degrees Fahr. in an Atmosphere of 34-35 Per Cent CO.

(G)   A micrographic inspection at the sectioned surfaces of a molybdenum-tungsten steel treated at 2240 degrees Fahr. in an atmosphere of 34 to 35 per cent CO is of some interest as it shows pictorially (1) the carbon absorption which results in treatment and (2) the tendency to a diminishing carbon at the surface.

The chief point of variance between the molybdenum-tungsten steel treated in this high CO atmosphere and a 0.70 per cent carbon 18-4-1 steel treated in either a 10 per cent CO or a 2.5 per cent O_2 atmosphere is to be found in the more extensive structural difference of the carburized zone as revealed by etching in 4 per cent Nital. The difference is one of degree as some massive carbide may be found near the surface of treated 18-4-1 steel.

The steel used in this test showed the following analysis:

C	Mn	Si	Cr	W	Mo	V
0.75	0.28	0.23	3.92	1.37	8.82	1.17

Samples approximately $\frac{13}{32}$ x 1 x 4 inches long after hardening from 2240 degrees Fahr. were examined at the sectioned surfaces in the as-quenched condition and after drawing at 1025 degrees Fahr. for one hour. A sample after lead annealing at 1400 degrees Fahr. for ten minutes showed on the surface cut 0.000 to 0.003 inch a carbon content of 1.00 per cent.

Fig. 22 (Sample No. 2) illustrates the as-quenched structure at the edge of the sectioned hardened surface. There is a some-

Table XII

Comparison of the Rockwell and Microhardness (Knoop) Numbers of the Hardened and Drawn Surface and the Surface of a Section of 0.75 Carbon-Molybdenum-Tungsten High Speed Steel Heated for Hardening in an Atmosphere of 34-35 Per Cent CO at 2240 Degrees Fahr.

Sample No.	Test Location	Draw Deg. F.	Rockwell "A"	Rockwell "C"	Microhardness Numbers (Loads in Grams) 2000	1000	500	200
4	.0006" below hardened surface	1050 1 hour	85.1	67.1	646	640	672	698
	Surface of section		84.2	66.1	772	787	795	799

what distintegrated ferritic zone about 0.0005 inch thick at the surface.

Fig. 23 (Sample No. 4) shows the edge structure after drawing 1025 degrees Fahr. for one hour. At the surface a darker structure for about 0.0005 inch would appear indicative of a lower carbon content than in the lighter colored zone beneath.

Fig. 24 shows sample No. 4 after etching in boiling sodium picrate. Etching indicates a ferritic exterior for about 0.0005 inch and beneath this a massive carbide zone for approximately 0.002 inch.

In Table XII the austenitic character of the massive carbide zone of sample No. 4 is shown by the microhardness numbers obtained 0.0006 inch below the treated surface. On a surface which was originally about one half the distance from the surface to the center of the same sample the microhardness values are considerably higher.

DISCUSSION OF RESULTS

An effort has been made to form some picture of the probable disposition of the carbon in approximately the first 0.012 inch of "surface" of 0.60, 0.70 and 0.93 per cent carbon 18-4-1 round high speed steel hardened from 2350 degrees Fahr. using an atmosphere of 10 per cent CO or an atmosphere of 2.5 per cent O_2. Due to the considerable number of variables natural to such an investigation, specific data is of genuine consequence only insofar as it fits the pattern as a whole.

The surface attack on high speed steel treated at 2350 degrees Fahr. appears to be of the nature of pit and intergranular attack. Consequently starting with such a surface some variant would be introduced due to a lack of any means for gaging freedom of the surface from extraneous adhering or entrapped carbon. The samples quenched from the 1550 degrees Fahr. furnace into oil indicate on the carbon analyses a higher carbon by 0.01 to 0.03 per cent in the cut 0.000 to 0.003 inch as compared to the carbon of the original samples. Due to the greater sacrifice of surface in "annealing" and sandblast cleaning the samples hardened from the high temperature furnace, the possible variant may be less than this. However, with the more extensive pitting due to the 2350 degrees Fahr. heat it may even be higher. That the surface zone is relatively high in carbon compared to the carbon of the original steel is indisputable in those samples from which 0.002 inch of surface was removed by grinding prior to obtaining the chips for analysis.

The steeper slopes of the curves obtained on the 0.60 and 0.70 per cent carbon high speed steels treated in the 10 per cent CO atmosphere compared to those of the 2.5 per cent O_2 atmosphere may indicate some or more surface heating in the case of the O_2 samples. However, surface phenomena at the 2350 degrees Fahr. temperature are not well understood and the greater diffusion of carbon in the O_2 atmosphere may be due to a greater chemical activity at the steel

surface by this atmosphere. It is believed that much of the carbon absorbed by the resulting treated surface is derived from the carbon of the exterior which is oxidized or scaled or fluxed off (if borax is used) during the sojourn in the 2350 degrees Fahr. furnace. The oxidized carbon would be nascent and in almost immediate contact with what would later be the final treated surface. Also some soot from the atmospheres may deposit on the steel surfaces and this be absorbed by the steel.

The curves (Figs. 7 and 8) showing the probable distribution of the carbon in treated round 18-4-1 "surfaces" indicate a diminishing carbon 0.001 inch below the surface to the surface of the 2.5 per cent O_2 atmosphere samples. The points are extremely meager as evidence. However, we believe the suggestion of a lower carbon content at the surface is indicative of the true condition which exists at the finished treated surface. The micrographs of the 18-4-1 steels and of the molybdenum-tungsten steel indicate this to be true.

In the carburization which appears natural to the heat treatment of 18-4-1 steel at 2350 degrees Fahr. the absorbed carbon is practically all matrix or solution carbon. Carbides as visible entities do not appear to form. Consequently there is usually no extensive structural change evident when carbon is absorbed at the surface of 18-4-1 high speed steel as a result of heat treatment. There are some color differences as well as some small amount of massive carbide near the surface which may be brought out by etching the section of the surface zone. Such measurement is qualitative rather than quantitative. The molybdenum-tungsten steel treated in the 34 to 35 per cent CO atmosphere shows pictorially the structural change due to the absorbed surface carbon. Even here the extent of the carburization is not entirely revealed by microscopic examination. The average scaling (or loss in diameter) in hardening the 0.93 per cent carbon 18-4-1 steel samples compared with the 0.70 per cent carbon 18-4-1 steel samples was of the order 6.5 to 1 when the 10 per cent CO atmosphere was used and 2 to 1 when the 2.5 per cent O_2 atmosphere was used. The average loss in diameter of the 0.70 per cent carbon 18-4-1 samples hardened from the 10 per cent CO atmosphere compared to the 2.5 per cent O_2 atmosphere was of the order of 1 to 3. There was practically no difference in the average loss in diameter of the 0.93 per cent carbon 18-4-1 steel whether treated in the 10 per cent CO or in the 2.5 per cent O_2 atmosphere. No direct comparison can be made with the 0.60 per

cent carbon 18-4-1 steel and the other two steels as the samples were smaller in diameter.

Under the conditions imposed in this investigation 0.95 per cent carbon would appear to approximate a "surface" percentage quite rapidly attained when 18-4-1 steel is treated at 2350 degrees Fahr. After this the carbon pick up decreases or the carbon diminishes at the surface. The extent of the scaling (or loss in diameter) would appear dependent on how rapidly the surface attained to about 0.95 per cent carbon. In the case of the 0.93 per cent carbon 18-4-1 steel this percentage would be attained almost as quickly in the 10 per cent CO atmosphere as in the 2.5 per cent O_2 atmosphere. It would appear that this high carbon surface is subject to a more rapid oxidation at the 2350 degrees Fahr. temperature.

The results of "surface" carbon determinations in Table VII would suggest that we might more readily approach neutrality as regards the carbon content of treated 18-4-1 high speed steel surfaces by designing a steel for our atmospheres rather than designing our atmospheres for our steels.

The microhardness tests by showing a greater hardness near the surface as compared to 0.0154 inch below the surface would appear to confirm the carburization which occurs in heat treatment.

The rates at which the chemical reactions (with respect to carbon) occur at the surface of the high speed steel treated at 2350 degrees Fahr. are extremely rapid. Consequently because of this high reaction rate and the presence of a considerable number of variables such as straightness of the machined samples, etc., precise duplication of the carbon results on specimens treated at different times are difficult to obtain. However, the absorption of carbon by the treated surfaces of all the high speed steels investigated do show the invariable pattern of carbon absorption.

stin of the Metallurgical Department of the Vanadium-Alloys Steel Company prepared most of the photomicrographs and Mr. Myron Frost, Chief Chemist for the same company made the complete analyses of the several heats of steel. Mr. Jean Rife, Chemist for the Landis Machine Company, supervised the machining of the samples and made the numerous carbon determinations.

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DISCUSSION

Written Discussion: By W. A. Schlegel, metallurgical department, The Carpenter Steel Company, Reading, Pa.

A careful study of the paper submitted by Mr. Morrison reveals that the work was thoughtfully planned and painstakingly and thoroughly executed. We owe Mr. Morrison our congratulations and thanks for this worthwhile contribution to our meagre knowledge of the reaction of furnace atmosphere with the steel during its heat treatment. Those of us connected with the manufacture of steel are especially grateful for this contribution to the literature by a consumer who has taken such an active interest in the problems of heat treating.

During the past few years we have made extensive studies of the effects of different furnace atmospheres during the heat treatment of 18-4-1 high speed steel. It is gratifying to find that Mr. Morrison and ourselves are in agreement. As an example the following are cited in confirmation of his results:

During heat treatment, regardless of the type of atmosphere used as commonly encountered with the usual heat treating furnaces, some oxidation takes place. In each instance, it was found that oxidation penetrates deeper in the austenitic grain boundaries than into the matrix proper. This may be described as intergranular oxidation. Mr. Morrison has observed the same characteristics which he shows in Fig. 3 .

Another point of agreement is shown in Fig. 4 where it is apparent that the rate of carburization caused by atmospheres is greater for short-time treating periods, and as the time is increased, the amount of carburization decreases. We have carried this slightly further and found that eventually decarburization will occur with the same atmosphere which will cause carburization with short heating cycles.

A third point of agreement is shown by Figs. 5, 6, 7, 8 and 9. These curves show that lean CO atmospheres, or oxidizing atmospheres, carburize more readily than rich CO atmospheres. Our results confirm these findings, and this is especially true when the heating cycles are increased up to 30 minutes at heat.

The excellent photomicrographs shown by Mr. Morrison were obtained after double tempering. It is known that the structures obtained due to the carburization consist of a mixture of martensite and austenite. Since these photomicrographs were taken after double drawing, the discusser would like to ask Mr. Morrison what the structure of the edge showed after a single draw. It has been our experience that double drawing is necessary to break down all of the retained austenite. In the event that the carburization is excessive, triple drawing is required to decompose all of the remaining austenite.

In Table X Mr. Morrison shows the effect of hardness due to carburization as determined by a microhardness tester. These results show that the hardness determinations were made after double drawing. It would be interesting to learn what the hardness results on the carburized case would have been in the as-quenched condition. In addition, the removal of successive small increments of the material in the as-hardened condition and the determinations of the hardness after each removal would serve as a gage as to the amount of retained austenite and possibly the degree of carburization.

Another interesting study along the same line would consist of determining the hardness at some specific level below the surface, preferably 0.001 or 0.002 inch, and determining the hardness as hardened, as tempered at 1050 degrees Fahr. (565 degrees Cent.), and as retempered as 1050 degrees Fahr. The tempering procedure could be continued until consistent results would be obtained. This would be a valuable addition to the theory that re-tempering is beneficial. Possibly Mr. Morrison has some information along these lines which he might care to contribute at this time.

A very important observation made by Mr. Morrison as shown in Table VIII is the degree of distortion caused by carburization. The marked difference in the degree of distortion between the 0.70 per cent carbon and the 0.90 per cent carbon 18-4-1 high speed steel is assigned to the excess carburization of the 0.70 per cent carbon material while the 0.90 per cent carbon steel did not show any appreciable gain in carbon content.

In closing this discussion the writer again wants to congratulate Mr. Morrison for this splendid paper.

Written Discussion: By W. R. Frazer, metallurgist, Union Twist Drill Co., Athol, Mass.

Mr. Morrison has conducted a very interesting investigation of a subject that is extremely important to all high speed tool hardeners. We are sorry that he did not carry on his original plan to compare the action of 1.5 per cent carbon monoxide with the higher concentration of this gas, as we are not aware of any high speed tools being deliberately hardened in an oxidizing atmosphere. Experience has shown that an oxidizing atmosphere gives less control over size change and produces a rougher surface than a reducing atmosphere, probably due to the greater amount of scaling, as shown by Morrison, and also the more rapid heating and possible overheating due to oxidation.

An interesting feature of this study is the observation that an oxidizing atmosphere carburizes high speed steel more effectively than a reducing atmosphere. We note that similar test samples were given the same time-temperature hardening cycle in both atmospheres. There is a decided difference in the

rate of heating of a piece of high speed steel in a 2.5 per cent oxygen atmosphere and one containing 10.0 per cent carbon monoxide, the former being very much faster. We can expect, therefore, that the samples heated in the oxidizing atmosphere were up to heat sooner and soaked at heat longer than those pieces heated in the reducing gases. Chemical reactions take place very rapidly at elevated temperatures, so the element of time is very important and if carburization is taking place, the sample at heat longer would show deeper penetration. We would not expect a duplication of these results if the test pieces had been removed from the oxidizing atmosphere immediately after they had reached the furnace heat.

The Knoop hardness values listed in Table X show greater hardness near the surface of the test pieces than 0.0154 inch below the surface, indicating the increased hardenability of the higher carbon zone. We note that these samples were all double tempered and would like to ask Mr. Morrison if similar results are obtained before tempering and again after the first tempering. We are of the opinion that the higher carbon surface retains more austenite, which will show a lower hardness value as hardened and also as tempered once, but repeat tempering will cause this austenite to transform and show the higher hardness as reported. This should be particularly true of the 0.70 per cent carbon samples which show a layer of undecomposed austenite in Figs. 16 and 17 even after double tempering. Mr. Morrison's comments on this will be appreciated.

May we congratulate Mr. Morrison on his competent presentation of the data he obtained while studying this subject which is so important to all tool hardeners.

Written Discussion: By J. A. Nelson, metallurgist, Braeburn Alloy Steel Corporation, Braeburn, Pa.

The subject and data given in Mr. Morrison's paper will be of great interest to all tool manufacturers and particularly to those in charge of heat treating high speed steel tools.

Surface reactions occurring in the treatment of high speed steel have been very little understood and the information given in this paper will help considerably in clearing up some of the unexplained conditions.

We note Mr. Morrison's remarks concerning the time element, in which was stated that as the time is increased the carbon pick-up is less, and that by extended duration of time a decarburizing action occurs. This observation agrees with some experiences encountered in the hardening of cutters where it was found that large pieces requiring up to 15 minutes in the high temperature furnace showed a soft skin, whereas cutters held in the same atmosphere for only 7 minutes did not show this same surface condition. This work was carried on in an electric furnace with controlled atmosphere—7 per cent CO, at 2350 degrees Fahr. (1290 degrees Cent.) It may be of interest to remark that these same soft cutters when annealed, reformed and rehardened in a gas fired, semi-muffle furnace gave satisfactory results. We believe it would be of great interest to learn which atmosphere would be more suitable for long duration at temperature, and which for short duration.

Regarding the influence of shape and carbon pick-up, we ran into a peculiar condition several years ago, when it was found by microscopic examination

that hardened and tempered sample disks which had been drilled for chemical analysis and hardened in a gas furnace showed a coarse martensitic structure surrounding the hole, and that this condition was not noticeable on the outside surface of the disks. Without any further investigation, we at that time attributed this condition to overheating at these thinner sections, but it now appears that carburization may have been a factor.

Author's Reply

In regard to Mr. Schlegel's question as to what the edge structure would be after a single draw compared to a double draw, I take it he is referring to the high carbon austenitic zone. This zone on an 18-4-1 steel is deeper or thicker after a single draw than after a double draw. Upon repeated tempering there is a progressive decomposition of the high carbon retained austenite, for the most part, from the interior of the zone outward. These observations, we believe, are in agreement with those of Mr. Schlegel's.

The high carbon austenitic zone has been termed, perhaps injudiciously, massive carbide. The term massive carbide has been used descriptively rather than definitively. It is obviously not carbide in toto. The characteristics of this zone resemble in many respects those of carbide more than they resemble those of the familiar austenite. In the case of the high carbon exterior zone of the molybdenum-tungsten steel (illustrated in Fig. 23) it is extremely difficult to decompose this zone at the temperatures usually employed for the tempering of this steel. In fact the decomposition of this zone within a reasonable time can only be accomplished at a temperature considerably in excess of that which can safely be used for the tempering of the metal below this high carbon exterior zone.

Dr. Frazer's observation that the samples heated in the 2.5 per cent oxygen-containing atmosphere would heat faster than similar samples heated in a 10 per cent carbon monoxide atmosphere is indeed true. Dr. Frazer questions that the samples heated in the 2.5 per cent oxygen atmosphere would have shown as high a surface carbon had they been removed from the superheat immediately after reaching temperature. To answer this question adequately, curves similar to those of Fig. 4 derived from samples treated in an atmosphere containing 2.5 per cent oxygen would be necessary. Such curves superimposed on those of Fig. 4 would show the relative speeds of carbon absorption by the two atmospheres with increase in time. We agree with Dr. Frazer that the diffusion of the carbon in the samples heated in the 2.5 per cent O_2 atmosphere to a greater depth than in the 10 per cent CO samples is due to the surface of the samples treated in the 2.5 per cent O_2 atmosphere reaching the 2350-degree Fahr. (1290-degree Cent.) temperature earlier. At the surface or rather about 0.001 inch below the treated surface the indicated carbon content is higher for the samples treated in the 10 per cent CO atmosphere than for the 2.5 per cent O_2 atmosphere samples. It is believed that if the 2.5 per cent O_2 samples had been removed earlier from the superheat a somewhat higher surface carbon would have been found; also a lesser diffusion of the carbon. Due to the decarburizing action setting in earlier in the case of the 2.5 per cent O_2 samples, the carbon content at the surface is somewhat lower than that found in the 10 per cent CO samples.

We have done some work along the line suggested by Messrs. Frazer and Schlegel in that we have attempted to obtain the microhardness or Knoop numbers near the treated surface in the as-quenched, drawn and double drawn conditions.   The results show a considerable increase in hardness at the "surface" after a single draw compared to the as-quenched hardness, and another considerable increase after a second draw.   Tests were made on ¾-inch diameter by 1-inch long samples taken from 1-inch diameter bars made of 30-pound induction melts of a 0.74 per cent carbon and a 0.92 per cent carbon 18-4-1 steel.   Samples were polished on one end before hardening at 2350 degrees Fahr. (1290 degrees Cent.) in an atmosphere of 10 per cent CO. After obtaining the as-quenched hardness, specimens of each steel were cut with a nicking wheel to form two half moon sections by ½ inch long.   All half moon sections were drawn at 1050 degrees Fahr. (565 degrees Cent.) for 1½ hours; one half moon section of each steel was then redrawn at 1050 degrees Fahr. (565 degrees Cent.) for 1½ hours.   The table shows the relative hardness values obtained on the two steels as-quenched, after a single draw and after a double draw.   The depth of the test surface below that of the as-quenched surface is given in the table.   The hardness numbers of the

Comparison of the Rockwell "C" and Knoop Microhardness Numbers Near the Treated Surface of a 0.74 and 0.92 Per Cent Carbon 18-4-1 Steel in the As-Quenched, Drawn and Double Drawn Conditions.

Specimen No.	Carbon	Condition	Surface Tested (Depth below as-quenched surface ± .0002 inch)	Rockwell "C"	Microhardness Numbers (Loads in grams)			Remarks Impressions measured at 350X
					2000	1000	500	
1		As quenched	0.0007 inch	65.1	689	658	697	Impressions clean cut. All oxidized surface removed.
1B	0.74	Drawn 1050 degrees Fahr. 1½ hours	0.0012 inch	66.1	787	778	795	—do—
1C		Redrawn 1050 degrees Fahr. 1½ hours	0.0013 inch	65.7	841	877	894	—do—
1		As quenched	0.0011 inch	62.0	538	513	532	Impressions jagged. Oxidized surface not entirely removed.
1B	0.92	Drawn 1050 degrees Fahr. 1½ hours	0.0016 inch	67.0	728	689	712	Impressions clean cut. All oxidized surface removed.
1C		Redrawn 1050 degrees Fahr. 1½ hours	0.0017 inch	68.0	820	856	878	—do—

single and double drawn specimens are comparable in that they represent tests on surfaces at the same depth below the original treated surface.   In the case of the 0.92 per cent carbon 18-4-1 steel, removal of 0.0011 inch from the as-quenched surface did not entirely dispose of all the oxidized surface metal whereas 0.0007 inch did in the case of the 0.74 per cent carbon 18-4-1 steel. The as-quenched "surface" of the 0.92 per cent carbon 18-4-1 steel as tested

was of a crumbly nature and the Knoop impressions were jagged in contour. All specimens were lightly etched before making the Knoop impressions. Impressions near or in the carbide segregate were not used in the calculations of the hardness numbers.

We have attempted to obtain the microhardness numbers of the high carbon austenitic zone but so far with the specimens used we have not obtained a sufficient depth to permit of preparation, without too great a sacrifice of the zone in question; also we have not secured a high carbon austenitic zone sufficiently free from oxidation pits to preclude interference from this source. It would appear that oxidation is concomitant to actual carburization; also that the surface carburization and surface attack referred to in the paper are of the nature of a bombardment rather than an even attack. None of the 18-4-1 specimens reported on showed the high carbon austenitic zone as a sheathing; rather these zones appeared as more or less discontinuous areas.

Mr. Nelson's observation of the importance of the time element in the superheat as regards a soft or decarburized surface is quite generally known. In the case of relatively large masses an atmosphere lean in CO makes for a more rapid heating; also the scaling or oxidation loss is greater, for equivalent time, in a lean CO atmosphere compared to a rich CO atmosphere. This greater surface oxidation loss is probably more nearly compensatory to the decarburizing action which proceeds after a certain time in the superheat .

We would like to introduce here the microhardness values obtained by Dr. Peters and Mr. Knopp of the National Bureau of Standards on a series

Comparison of the Rockwell "A" and "C" and Knoop Microhardness Numbers Taken 0.002 Inch and 0.010 Inch or 0.015 Inch Below the Treated and Double Drawn Surfaces of 0.60, 0.70 and 0.75 Per Cent Carbon 18-4-1 Steel Specimens, Treated at 2350 Degrees Fahr. in (1) an Atmosphere of 10 Per Cent CO and (2) an Atmosphere of 2.5 Per Cent O_2. Specimens ⅝ Inch x ½ Inch x 1 Inch. Specimens Drawn 1050 Degrees Fahr. 1½ Hours, Redrawn 1080 Degrees Fahr. 1½ Hours.

Carbon and Specimen No.	2350 Degrees Fahr. Atmosphere	Surface Tested (Depth Below Treated Surface) Plus or Minus .0002 inch	Rockwell "A" Scale	Rockwell "C" Scale	Microhardness (Knoop) Nos. Loads in Grams 2000	1000	500	200	100
.60—1	10 per cent CO	.002 inch	83.6	63.0	777	800	816	826	800
		.010 inch	83.4	63.0	737	764	777	801	793
.60—2	2.5 per cent O_2	.002 inch	83.5	63.9	800	819	836	796	792
		.015 inch	83.3	63.3	745	787	784	804	816
.70—1	10 per cent CO	.002 inch	85.0	65.1	835	864	866	870	831
		.010 inch	84.2	65.0	780	809	819	808	808
.70—2	2.5 per cent O_2	.002 inch	84.1	65.0	828	870	843	879	908
		.015 inch	84.0	64.6	793	810	822	854	893
.745—1	10 per cent CO	.002 inch	84.8	65.3	812	860	858	870	849
		.010 inch	84.6	65.1	783	792	820	822	787
.76—2	2.5 per cent O_2	.002 inch	84.0	64.8	819	841	847	863	831
		.015 inch	83.9	64.6	782	798	802	788	806

of 0.60, 0.70 and 0.75 per cent carbon 18-4-1 steel specimens treated at 2350 degrees Fahr. (1290 degress Cent.) in (1) an atmosphere of 10 per cent CO and (2) an atmosphere of 2.5 per cent O_2. This was another set of specimens similar to that shown in Table X. These tests are more comprehensive as the hardness comparisons between the high carbon surface zone and the zone 0.010 or 0.015 inch below the treated surfaces have been made under indenter loads of 100, 200, 500, 1000 and 2000 grams.

DIMENSIONAL CHANGES ON HARDENING HIGH CHROMIUM TOOL STEELS

By Howard Scott and T. H. Gray

Abstract

Identifying and controlling the external factors affecting the dimensional changes that steel undergoes on hardening, it was possible to evaluate accurately the changes inherent in the steel itself. With the hardening practice found necessary, two important air hardening tool steels were carefully studied. While the results indicate clearly that with good hardening practice dimensional changes are reproducible and can be anticipated so that the hardened piece has the predetermined dimensions, the present paper has for its purpose not the description of the commercial aspects but presentation of very accurate measurements obtained under the closest possible conditions of control.

Introduction

ONE of the most baffling problems confronting the maker of precision tools is control of the dimensional changes which occur on hardening. Great expense may be incurred in correcting the apparently erratic movement or distortion that accompanies hardening. Often the tool maker capitulates and grinds all working surfaces after hardening. In view of recent developments in heat treatment equipment and practice, this seemed to be an avoidable expense. With encouragement from production departments, a laboratory investigation of this problem was undertaken.

Extraneous Factors

A wide variety of factors are blamed for the permanent dimensional changes occurring when steel is hardened and their inconsistencies. Obviously it is necessary to consider every possible factor and to find which contributes materially to the effects in question. The factors considered here are:

1. Segregation in the steel used.

2.   Incomplete hardening, particularly at surfaces.

3.   Distortion of piece in the furnace under its own weight.

4.   Metal loss from scaling.

5.   Release of pre-existing strain in the work due to heating for hardening.

6.   Temperature gradients during heating.

7.   Temperature gradients during quenching.

8.   Inherent increase in specific volume of steel on quench hardening.

9.   Directional effects in the inherent volume change.

10.   Decrease in specific volume of hardened steel on tempering to lower hardness.

Discussion of these factors in the preceding order will indicate how some were circumvented or eliminated in the present work isolating the others for detailed study.

The steel used in this investigation passed deep-etch acceptance tests so segregation, item (1), was no greater than must be expected in a commercial product. Experience with production hardening of complicated dies confirmed this conclusion indirectly because the die users reported consistent dimensional changes from the hardening practice developed.

Incomplete hardening, item (2), was eliminated as a factor by use of steels having ample hardenability to secure full hardening, at least 64 Rc, throughout the heaviest section. The two steels used were of the following approximate composition:

Steel No.	Carbon	Manganese	Silicon	Chromium	Molybdenum	Vanadium
			Per Cent			
21	1.0	0.65	0.3	5.2	1.0	0.25
15	1.5	0.25	0.3	11.5	0.8	0.25

Steel No. 21 hardens fully in a solid block 2 x 6 x 10 inches when cooled from 1730 degrees Fahr. (945 degrees Cent.) in ammogas[1] and No. 15 in the form of a block 3 x 6 x 10 inches when so cooled from 1850 degrees Fahr. (1010 degrees Cent.).

Both in the furnace and cooling chamber the work was supported on a metal tray resting on a flat metal muffle floor. Thus distortion from handling with tongs while hot and from deformation due to irregular support was avoided, item (3). The practice and equip-

[1]Ammogas is a convenient term coined to designate completely dissociated and dried ammonia which, therefore, is composed of 75 per cent hydrogen and 25 per cent nitrogen.

ment used is fully described in a previous paper from this laboratory.[2]

Metal loss by oxidation, item (4), has been shown to be a factor in dimensional changes on hardening.[3] These changes are eliminated by use of the ammogas atmosphere which completely prevents oxidation. Thus four out of the ten possible variables in the hardening equation were eliminated from further consideration permitting concentration on the remaining six which required experimental study.

Residual stresses existing in the metal stock may easily cause movement during heating for hardening, item (5). Even before the critical range is reached, the steel becomes plastic allowing stresses to relax with accompanying movement. Such stresses are often introduced by cold straightening operations. In the present case however they were certainly very small because the stock from which test pieces were cut was forged and annealed 2.25 x 12 inch plates. The test specimens were rectangular blocks 2 x 6 x 6 inches having a circular hole 5 inches in diameter bored through their center, a convenient shape for measurement and one very prone to distortion.

To determine the magnitude of dimensional changes due to stress release in a specific case, dimensions were measured before and after an anneal for 1 hour just below the critical range. The annealing temperature was arbitrarily chosen as that giving minimum hardness on the tempering curve. It is recorded in Table I in the "preheat" column.

Dimensions were measured at the several points and directions indicated in Table I, the changes from those as annealed being recorded. Vernier micrometers were used and read to the nearest 0.1 thousandths of an inch under carefully controlled conditions. Precision gage blocks were used as reference standards and the micrometers so handled as to avoid heating by the hands.

The changes observed in both steels during the stress relieving anneal were negligible being in most cases within the limits of experimental error and none over 0.15 thousandths of an inch per inch. Evidently then dimensional changes due to release of residual stresses are of negligible magnitude in fully annealed tool steel that has not been straightened or otherwise cold deformed.

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[2]J. R. Gier and Howard Scott, "Bright Hardening of Tool Steels Without Decarburization or Distortion," TRANSACTIONS, American Society for Metals, Vol. 28, 1940, p. 671.

[3]Howard Scott, "Dimensional Changes Accompanying the Phenomena of Tempering and Aging Tool Steels," TRANSACTIONS, American Society for Steel Treating, Vol. 9, 1926, p. 277.

Table I
Change in Dimensions on Heat Treatment of Distortion Test Specimens in Thousandths Per Inch

Steel No.	Preheat 1 hour	Hardening Temperature 1 hour	Cooling Medium	Dimensional Changes					
				Internal Diameter			External Dimensions		
				Long.[1]	Trans.[2]	Diag.[3]	Long.[1]	Trans.[2]	Thick.[4]
21	1350° F.	None	Gas	+0.13	+0.02	+0.05	+0.13	+0.05	+0.02
21	None	1735° F.	Gas	+1.20	+1.30	+1.20	+1.20	+1.28	+1.15
21	1350° F.	1735° F.	Gas	+0.58	+1.68	+1.23	+0.98	+1.35	+0.50
21	1350° F.	1735° F.	Oil*	+0.80	+0.70	−0.70	+0.99	+0.85	+1.05
15	1380° F.	None	Gas	−0.15	+0.13	0.00	−0.08	+0.13	0.00
15	None	1885° F.	Gas	+0.80	−0.60	+0.15	+0.65	−0.64	−1.15
15	1350° F.	1885° F.	Gas	+1.34	−0.20	+0.56	+1.03	−0.21	−0.60
15	1350° F.	1885° F.	Oil	+1.30	−0.88	−0.32	+0.81	−0.64	−1.10

*This specimen cracked badly after quenching.
[1]Dimensions measured in direction of hot working.
[2]Dimensions measured perpendicular to direction of hot working.
[3]Dimensions measured at 45 degrees to direction of hot working.
[4]Change in thickness measured at corners.

Preheating of dies prior to exposing them to the hardening temperature is a common practice. Presumably it is done to avoid deformation from the thermal gradients incident to heating, item (6). Not knowing how important a factor this is, specimens were hardened with and without preheating. If temperature gradients during heating are an important factor, there should be a notable difference between the changes with preheat and those without because much higher gradients develop in the latter case. Cooling conditions were of course identical. This experiment was performed with the same specimens as used for determining the effect of a stress relieving anneal. One specimen of each steel was introduced directly into the furnace while it was at the hardening temperature for that steel. With preheating the cold specimens were introduced directly into the preheat zone while it was at the preheat temperature of 1350 degrees Fahr. (730 degrees Cent.).

The results of this experiment are given in Table I and illustrated diagrammatically in Fig. 1. Surprisingly the results are more consistent without the preheat than with it. Omission of the preheat certainly causes no distortion in No. 21 steel. In the case of No. 15 steel there is more longitudinal and less transverse change without the preheat, but the algebraic sum, an index of total movement, is unchanged.

In the last experiments temperature gradients during quenching, item (7), were eliminated as a variable simply by holding them constant. Moreover, they are very low with cooling in a gaseous media and could not then be an important factor. To show the relative

magnitudes of gradients for gas, oil and water quenching, the temperature difference between center and surface of a cylinder was calculated for comparable conditions with the results given in Table II. Here it is seen that the advantage gained by going from oil to air quenching is much greater than that from water to oil.
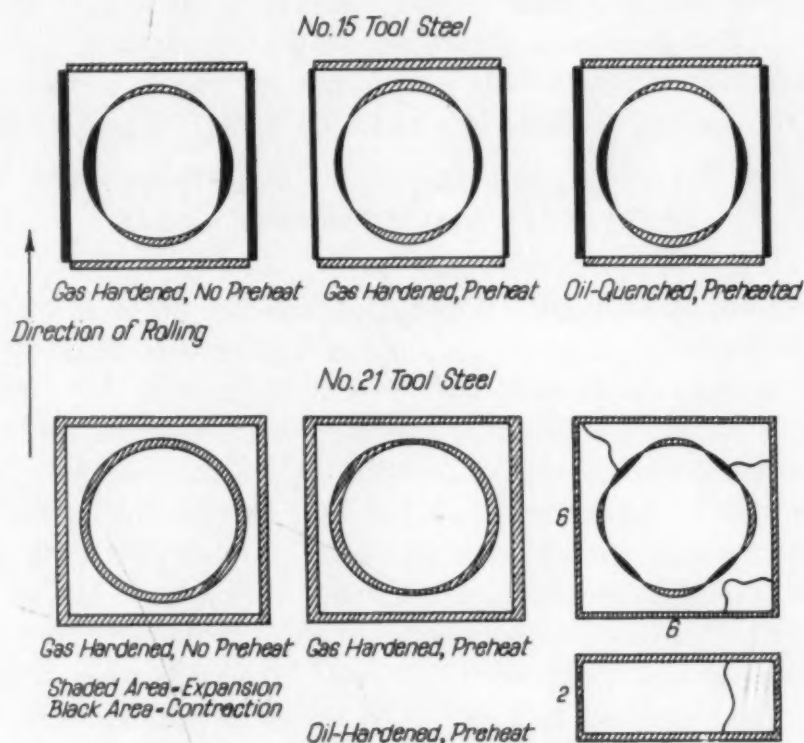

Fig. 1—Dimensional Changes in Distortion Test Specimens of No. 15 and No. 21 Steels Exaggerated for Illustration. Test Specimens 6 Inches Square, 2 Inches Thick with Central Hole 5 Inches Diameter.

To attempt to show the effects of temperature gradients quantitatively would be unprofitable because of the great complexity of the problem. After all, we know that temperature gradients of high magnitude cause severe distortion which is a complicated function of the shape of the piece and that the only way to avoid this type of distortion is to use a slower quenching medium. That cooling in a gaseous media, the slowest practicable quench, eliminates temperature gradients as an important factor in dimensional changes is attested by its extensive successful use for production hardening by Westinghouse plants.

Measurements were also made on a Navy distortion test specimen[4] of No. 15 steel before and after hardening to 65.5 Rc. The

[4]Described in Metals Handbook, p. 664, 1939 Edition.

alignment of the specimen in the bar stock from which it was cut is not known, but can be deduced from the changes recorded. In the direction defined by the center of the disk and of the hole bored in it, the change was a shrinkage of 0.14 thousandths per inch. Perpendicular to that direction there was an expansion of 1.06 thousandths per inch indicating clearly that this is the rolling direction. Thickness change was a shrinkage of 0.87 thousandths per inch. That the notch closed 0.0020 inch is not surprising in view of the marked directional characteristic to which it may be attributed.

Inherent Characteristics of Steels

It is now clear that the dimensional changes recorded in Table I for hardening treatments with cooling in a gaseous media are entirely due to the inherent increase in specific volume, item (8), which all tool steels undergo when hardened by quenching. In the case of No. 21 steel this change amounts to 0.30 per cent normally, but in No. 15 steel is much less, 0.11 per cent despite the fact that its change in the longitudinal direction may be as high as that of No. 21. Scott[5] also measured the inherent volume increment in several steels by means of density measurement which eliminated scaling effects and his results are compared with those observed here in Table III.

An important fact is brought out by Table III, namely that the volume increment diminishes materially with increase in chromium content. While this is a highly desirable effect it is accompanied by an objectionable phenomenon at high chromium concentrations, namely, appearance of a pronounced directional effect, item (9). This effect is pronounced in the 11 per cent chromium steel, No. 15, but absent in the 5 per cent steel, No. 21. Thus No. 21 is the more desirable steel from the viewpoint of dimensional changes because they are uniform in all directions and predictable, though the net volume change is considerably larger.

Steel No. 15 would be preferable to No. 21 could it be so treated as to eliminate the directional effect. One experiment made in this direction was to anneal a test specimen at 1150 degrees Cent. (2100 degrees Fahr.) for 1 hour prior to hardening and measurement of dimensional change. The directional effect was presumed to be due to carbide banding and later verified. If then the carbides could be

[5]Loc. Cit.

Table II

Calculated Maximum Temperature Difference Between Center and Surface of Steel Cylinders During Quenching from 930 Degrees Cent. (1705 Degrees Fahr.) to Medium at 30 Degrees Cent.

Cylinder Diameter in Inches	Temperature Difference in Degrees Cent. for Quench in		
	Water	Oil	Air
4	846	567	55
2	801	414	30
1	734	277	16
0.5	617	171	9

dissolved and reprecipitated in globular form the effect should be diminished. The specimen used was a solid cylinder 1 inch in diameter and 2 inches long. Its change in length was +0.9 and in diameter —0.7 thousandths per inch which is close to the change observed on No. 15 without preheat.

Evidently a more drastic treatment is necessary to obtain uniformity of the changes in No. 15 steel. Apparently the only feasible means for breaking up carbide banding is to upset forge. An experimental effort to do this was performed through the co-operation of J. S. Billingsley of the Crucible Steel Company of America. A plate 2.25 x 12 x 24 inches was forged by upsetting a 9 x 9 inch billet so that the original banding in the billet was partially reversed. A specimen 2 x 4 x 6 inches was cut from this plate and hardened in the standard manner. Its dimensional change was +0.45 thousandths per inch in the direction of plate length and +0.25 in the transverse direction. The upset forging practice therefore conferred a marked improvement in the changes though it did not completely eliminate the directional effect. In fact the changes are the lowest yet observed in any steel hardened throughout to 64-66 Rc.

Evidence that carbide segregation is the dominant factor causing variation in the inherent dimensional changes with direction relative to hot working is furnished by the microstructure of the upset bar,

Table III

Increase in Volume of Tool Steels on Hardening

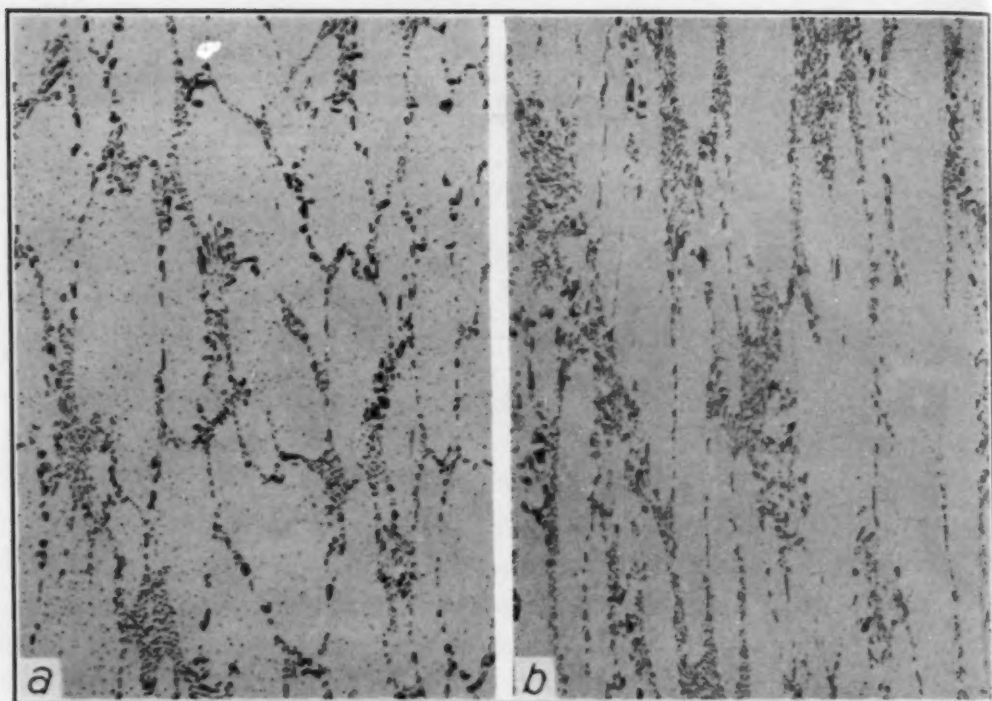| Carbon	Manganese	Chromium	Molybdenum	Tungsten	Vanadium	Heat Treatment Degrees Fahr.	Increment in Per Cent of Initial Volume
0.9	1.1	0.5	...	0.5	...	1440—to water	0.69
1.0	0.2	1.3	...	...	...	1470—to water	0.52
1.0	0.6	5.2	1.0	...	0.2	1730—to ammogas	0.30
1.5	0.2	11.5	0.8	...	0.2	1850—to ammogas	0.11

Fig. 2—Microstructure of Upset Forged Plate Compared with that of Normally Forged Plate of No. 15 Steel. Plane of Microsection Parallel to Length-Width Plane of Plate; Length of Plate Vertical. Magnification × 50.

Fig. 2a—Upset Forged Plate. Dimensional Change on Hardening +0.45 Thousandths Per Inch in Longitudinal Direction, +0.25 in Transverse Direction.

Fig. 2b—Normally Forged Plate. Dimensional Change on Hardening +1.0 Thousandths Per Inch in Longitudinal Direction, −0.2 in Transverse Direction.

Fig. 2a, which is compared with that of a normally forged plate, Fig. 2b. It is seen that a decided longitudinal banding is accompanied by a large longitudinal movement and the elliptical structure produced by upsetting is accompanied by only a small longitudinal movement relative to transverse.

So far we have considered only the net change from the annealed to the as-quenched condition, quenching being so performed that a hardness of 64 to 66 Rc is obtained throughout the body. If the quenching temperature be either raised or lowered materially, lower hardness is obtained with a smaller volume change. In the former case it is due to retention of austenite which is much denser than martensite and in the latter to a lower concentration of carbon in solid solution. With the high quenching temperature exceptionally high hardness values can be obtained in both No. 15 and No. 21 steels by cooling them to the temperature of dry ice, but the overall volume increment is correspondingly larger.

Tempering reduces hardness and since it is accompanied by pre-

cipitation of carbide should result in a shrinkage approaching the annealed density when carbide precipitation is complete, item (10). Figs. 3 and 4 show clearly that this is the case. The observations plotted were made largely on test cylinders 1 inch diameter and 2 inches long, but many observations have been made on other shapes including production dies, which are also plotted. It is a very striking
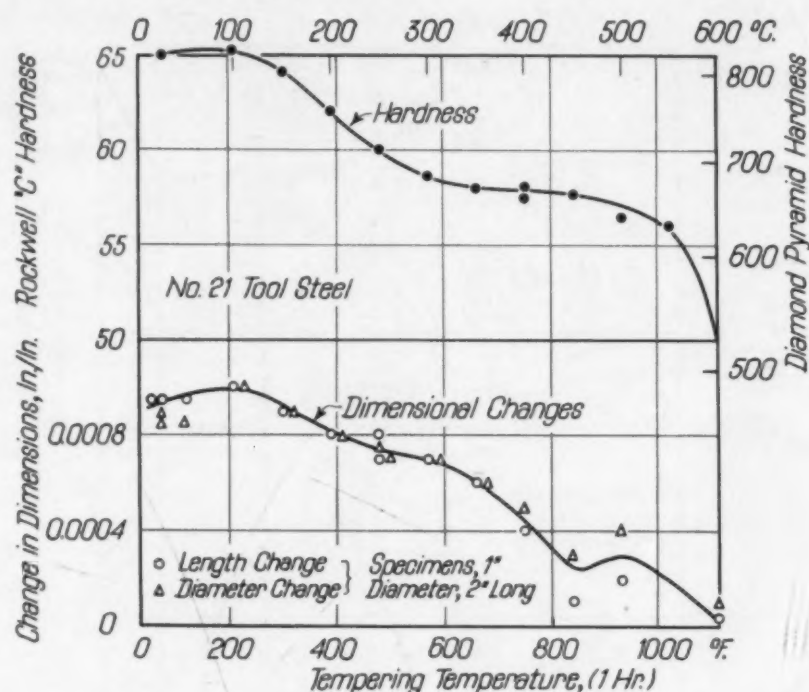

Fig. 3—Hardness and Dimensional Changes on Tempering No. 21 Steel After an Ammogas Quench from 1730 Degrees Fahr. (945 Degrees Cent.).

fact that the unit changes either on quenching or tempering do not depend on the shape of the work, at least when uniformly hardened without severe temperature gradients. This means that it is quite possible to correct for dimensional changes when a tool or die is made obtaining the desired final dimensions provided that it is hardened with all of the essential precautions enumerated here.

Shrinkage on tempering also has no directional component at least in No. 21 steel. Thus on tempering at 1150 degrees Fahr. (620 degrees Cent.) (515 DPH or 49 Rc) it returns to its original dimensions or within a few ten-thousandths thereof. Tool steels are seldom used at so low a hardness so little practical use can be made of this fact.

Neither is there evidence of a directional tempering effect in the

early stages of tempering No. 15 steel. As a result the net change
is a greater shrinkage between annealed and tempered state in the
transverse direction with reduced expansion in the longitudinal. Thus
the directional effect, indicated by difference between longitudinal
and transverse change, does not diminish in magnitude much on tem-
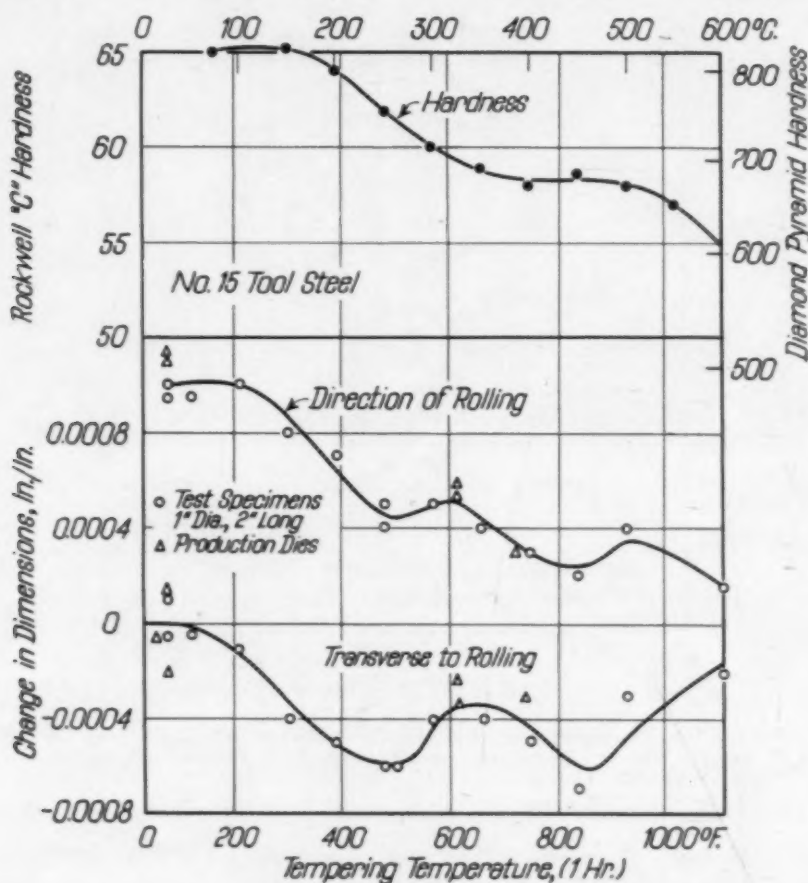pering up to about 800 degrees Fahr. (425 degrees Cent.). From

Fig. 4—Hardness and Dimensional Changes on Tempering No. 15
Steel After an Ammogas Quench from 1850 Degrees Fahr. (1010
Degrees Cent.).

that temperature on, this steel expands in the transverse direction and
contracts in the longitudinal approaching the original dimensions with
tempering at 1150 degrees Fahr. (620 degrees Cent.). Apparently
then there are two distinct stages in the tempering reaction, one in
which the contraction is uniform in all directions and another in
which the directional component is annulled, a phenomena deserving
further attention.

Though the volume increment on hardening tool steels diminishes
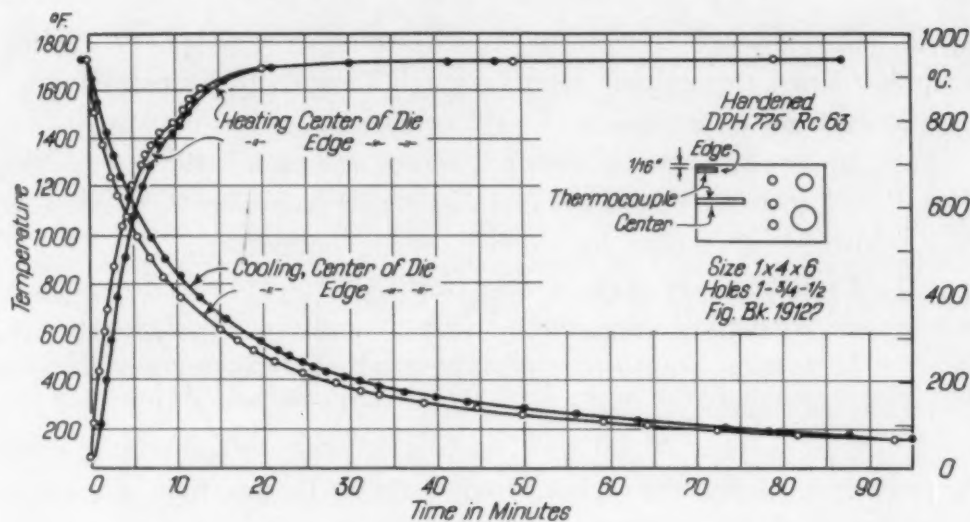materially with chromium content, inherent dimensional changes in

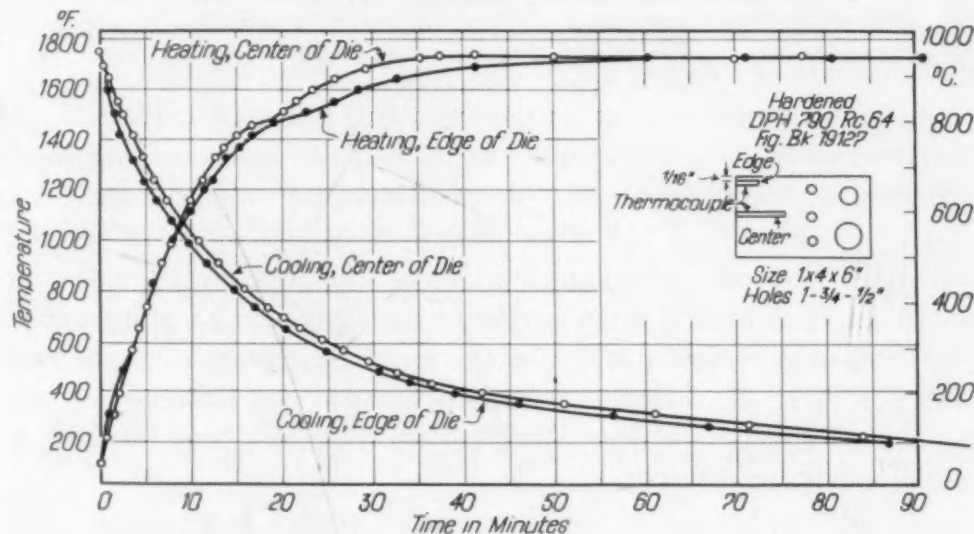Fig. 5—Heating and Cooling Curves of No. 21 T.S. Die Steel in Still Air.

Fig. 6—Heating and Cooling Curves of No. 21 T.S. Die Steel in Ammogas.

the high chromium steels are still of a magnitude very important to makers of precision dies and tools. So far as is now known, these changes cannot be completely eliminated without accepting either a much lower working hardness or hardening of the surface only to the desired value.

SUMMARY AND CONCLUSIONS

Eliminating seven possible causes of distortion and erratic dimensional changes on hardening, the dimensional changes inherent in the steel itself were studied in some detail. Attention was confined to two distinctive air hardening die steels which were heated for harden-

ing in and quenched by cooling in an ammogas atmosphere.   These steels hardened throughout to at least 64 Rc with this treatment.

From these experiments the following conclusions may be drawn:

1.   Movement due to residual stress relaxation during heating is negligible in work machined from annealed bars that have not been cold deformed other than by normal cutting operations.

2.   Preheating is unnecessary for the control of dimensional changes on hardening work even when of rather complicated shape.

3.   Distortion due to temperature gradients during quenching is effectively eliminated by "gas quenching."

4.   The volume of all tool steels increases on hardening when the piece is hardened throughout to at least 64 Rc and may amount to 0.7 per cent of the volume as annealed.

5.   The volume increment resulting from hardening diminishes rapidly with increasing chromium content, being only 0.1 per cent in an 11 per cent chromium steel.

6.   The movement on hardening a 1.0 per cent carbon, 5 per cent chromium steel is the same in all directions, amounting to an expansion of approximately 1.0 thousandths of an inch per inch.

7.   Practically all of the movement on hardening a 1.5 per cent carbon, 11 per cent chromium steel is in the direction of extension during hot working and equal to that of the 5 per cent chromium steel.

8.   The directional effect in the movement of the 11 per cent chromium steel was found to correlate with carbide banding.   It can be greatly reduced by upset forging which causes the carbides to assume a cellular pattern.

9.   Size and shape of test piece or of bar stock from which specimens were cut had no substantial effect on the dimensional changes observed.

10.   With the steels and hardening practice used here, dimensional changes are predictable even in complicated dies so that by compensating changes in dimensions before hardening, the desired final dimensions can be secured within commercial tolerances.

DISCUSSION

Written Discussion: By Sam Tour, consultant, C. I. Hayes, Inc., Providence, R. I.

The authors of this paper have shown that water quenching and oil quenching of steels results in larger changes of dimension than occur in air hardening. This fact has been one of the strong arguments for the use of air hardening

steels throughout the many years. They have shown that the dimensional changes on air hardening of a 1 per cent carbon, 5 per cent chromium, 1 per cent molybdenum steel are greater than the dimensional changes which occur on air hardening of a 1.5 per cent carbon, 11.5 per cent chromium, 0.8 per cent molybdenum steel. This also is a well recognized phenomenon.

That air hardening steels suffer some oxidation when cooled in the air is common knowledge. That this oxidation can be largely if not totally eliminated by cooling these air hardening steels in a nonoxidizing atmosphere is also common knowledge. This was demonstrated by one furnace manufacturer at the Metal Exposition some two years ago.

Although the authors of this paper make repeated reference to "Ammogas", they have not shown that Ammogas in itself offers anything more than a nonoxidizing atmosphere for the heating and cooling of so-called air hardening steels. It is quite obvious that similar results would be obtained in any other type of nonoxidizing atmosphere.

The conclusions drawn by the authors from the experiments described in their paper are entirely too broad and unwarranted. Specifically, conclusions No. 1, No. 2 and No. 10 should be limited to air hardening steels only. The authors have no proof that these remarks apply to other types of steel. Conclusion No. 3 should be limited to air hardening steels and the term "gas quenching" should be eliminated from that conclusion. The proper conclusion is the old accepted and obvious fact that "distortion due to temperature gradients during quenching is effectively reduced by reducing the temperature gradient and this is possible with air hardening steels."

Written Discussion: By S. C. Spalding, metallurgical engineer, The American Brass Company, Waterbury, Conn.

In looking over this paper I note that for hardening No. 21 steel a temperature of 1730 degrees Fahr. is used. Most of the suppliers' catalogs I have seen recommending this grade of steel suggest 1800 to 1850 degrees Fahr.

It is stated at another point in the paper that any change in hardening temperature, either up or down from that which the authors selected, would very materially affect the dimensional changes. Is there any particular reason why this steel should be hardened at temperatures so much below those recommended? If it had been hardened at 1800 or 1850 degrees Fahr. would it show up more favorably in comparison with No. 15 steel in regard to volume change than it would in the results that are indicated in the paper? As I judge the paper, the authors did not find very much difference between the two as regards volume changes, but, in general, the No. 21 steel had a little greater volume change than the No. 15, although it was equal in all directions.

What I am interested in is if the No. 21 steel had been hardened at the recommended higher temperatures, whether its comparison with the No. 15 steel would be less favorable.

Oral Discussion

L. C. Conradi:[a] My questions have to do with corrections on dimensions and measurements.

On page 505 the authors say, "Vernier micrometers were used and read to

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[a]Technical research director, International Business Machines Corp., Endicott, N. Y.

the nearest 0.1 thousandths". That means 0.0001 inch as I understand it. Is that correct? On the same page in the last paragraph I take it that 0.15 thousandths means 0.00015 inch.

On page 509 I think perhaps the authors made some corrections in presenting the paper. However, I take plus 0.45 inches to mean 0.00045 inch and the plus 0.25 inches to mean 0.00025 inch.

Finally under conclusion 6 on page 514 I assume 1.0 thousandths of an inch as mentioned there really means 0.001 inch.

These are the only questions I have and I would appreciate it if they could be corrected from the platform here.

Authors' Reply

Mr. Conradi's careful reading disclosed several errors which he has correctly corrected. The term "mil" was substituted in the presentation for "thousandths-of-an-inch."

Certainly it is generally known, as Mr. Tour states, that the volume change on hardening the 12 per cent chromium type steel is less than that of the 5 per cent type. That there is a prominent directional effect in the former, but not in the latter, however, is not so well known. Also, we are not aware of any previous evaluation of the inherent dimensional changes in these steels unobscured by oxidation effects, nor of any demonstration that there is no important size or shape factor in the movement of air hardening steels.

As Mr. Tour suggests it is no great achievement to harden air hardening steels without oxidation. The merit in the procedure and equipment used is that it eliminates decarburization and the unfortunate consequences of steep temperature gradients, as well as oxidation. This technique was first described at the preceding National Convention by Gier and Scott. A simple and easily obtained gas is used in a furnace specially designed for its efficient application.

Decarburization is a particularly disturbing factor in the heat treatment of air hardening tool steels because its progress is very rapid at the high hardening temperatures required and because wear resistance is lost where it is most needed. Of course surface decarburization has no important effect on the dimensional changes so its prevention was not essential in the present experiments. For this reason no mention was made in the text of the protection against decarburization afforded by "Ammogas" when properly protected against contamination.

That conclusion 10 should be limited to air hardening steels is conceded. In fact the opening clause so limits it. Conclusions 1 and 2 on the other hand are not so limited despite recognition of the implications. The difference between air hardening and less highly alloyed steels is only one of degree. Certainly little residual stress will exist in any steel after tempering for minimum hardness and, therefore, when preheated at the temperature so defined. It is concluded, therefore, that absence of movement at that preheat temperature is evidence of the absence of high initial residual stress whether the steel be of low or high alloy content. That preheating is unnecessary, conclusion 2, is simply based on the experience of hardening several hundred production die parts without cracking one and with reports of exceptional performance from

the shop. We have no objection to Mr. Tour's rewording of conclusion 3 and thank him for his pertinent comments.

Mr. Spalding notes that we use a considerably lower quenching temperature for No. 21 steel (5 per cent chromium) than recommended by suppliers and naturally wonders why. The reason is simply that maximum hardness is obtained at the quenching temperature of 1730 degrees Fahr. used or close thereto when change of surface carbon content is avoided. This is shown by the curves of Fig. A. Presumably to get maximum surface hardness in the presence
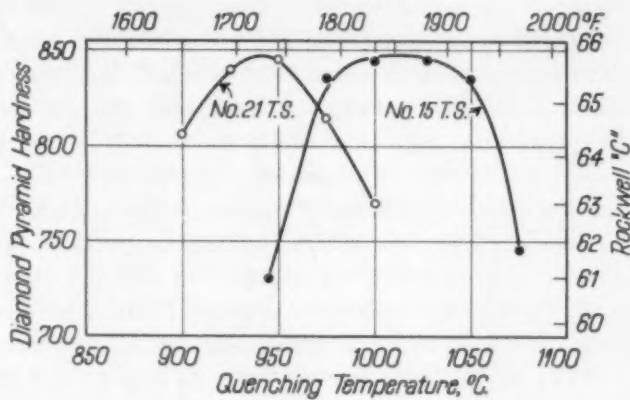
Fig. A

of decarburization a higher quenching temperature would have to be used. This may be the reason why higher temperatures have been recommended. Also they may be recommended simply to get greater penetration of hardening.

Table I

Dimensional Changes in 1-Inch Diameter Cylinders, 2 Inches Long Gas Quenched From a Carburizing Atmosphere

Quenching Temperature Degrees Cent.	Degrees Fahr.	Core Hardness DPH	Rc	Unit Change in: Length	Diameter	Volume Change Per Cent
			No. 21 Tool Steel			
900	1652	765	62.5	+1.2x10⁻³	+1.0x10⁻³	+0.32
940	1724	810	64.5	0.9	0.9	+0.27
1000	1832	715	60.5	0.8	0.7	+0.22
1000	1832	*805	64.0	2.7	2.7	+0.81
			No. 15 Tool Steel			
950	1742	760	62.5	+0.6	+0.7	+0.20
1010	1850	810	64.5	+0.7	+0.2	+0.11
1030	1886	†800	64.0	+0.9	−0.7	−0.05
1065	1949	720	60.5	+0.2	−1.1	−0.20

*After Cooling in Dry Ice.
†Previously Homogenized at 1150 Degrees Cent.

Mr. Spalding also requests information on the effect of quenching temperature on movement. What data we have on this factor is given in Table I. The cylindrical specimens were heated in a carburizing gas, but it is improbable that the carburized surface, depth 0.010 to 0.020 inch, affects the

dimensional changes materially because the volume of metal carburized is relatively small. In any event, the volume change in the carburized specimens is within 10 per cent of that uncarburized, Table III of paper.

It may now be seen that the volume change of both steels diminishes as the quenching temperature rises, an effect due to increase in content of retained austenite. That considerable austenite is present after quenching No. 21 from 1832 degrees Fahr. is shown by the hardness and volume increment on cooling in dry ice. While the initial volume change is slightly less for this quenching temperature than for 1724 degrees Fahr., which gives maximum hardness, it is difficult to see how any advantage is gained in view of the hardness loss. Of course, if tempered for secondary hardening, there will be a gain in hardness over the lower temperature, but then there will be an actual increase in volume on tempering because of the austenite transformation. This observation thus supports our contention that the best hardening temperature for these steels is that conferring maximum hardness as quenched.

Regarding comparative dimensional changes of these steels, it may be seen that the same relations exist between them at a higher quenching temperature as at that recommended. On the other hand it appears that the directional effect is absent in No. 15 steel quenched from 1742 degrees Fahr. This observation indicates a possible advantage in using a low quenching temperature on this steel when a hardness under 62 Rc is acceptable, but this single observation should be confirmed before accepting its indication as a fact.

We are indebted to Mr. Spalding for having stimulated a re-examination of our data which has brought out previously unrecognized facts.

QUANTITATIVE MEASUREMENT OF STRAIN HARDNESS IN AUSTENITIC MANGANESE STEEL

By D. Niconoff

Abstract

The strain hardening properties of austenitic manganese steel were determined by subjecting it to impacts of different intensities and measuring the surface hardness as well as the depth of its penetration. The maximum strain hardness of approximately Rockwell C-50 was found to be located not at the surface, but a short distance below it. Further down, the strain hardening effect gradually decreased until it disappeared completely. Under repeated impacts, this maximum strain hardness could be increased only slightly. With still greater number of the impacts delivered against already hardened surface of manganese steel, the hardness of the outer layers was observed to have actually decreased somewhat; but the depth of penetration of the strain hardness under such conditions was increased. It was also found that unexpectedly high work input was required for fully strain hardening austenitic manganese steel.

INTRODUCTION

DESPITE the wealth of literature available on the subject of austenitic manganese steel, there is relatively little known about the extent to which this steel is capable of hardening under the application of cold work. It is generally agreed that austenitic manganese steel possesses an unusually great tendency towards strain hardening, but very little systematic effort has been expended to ascertain accurately the relation between the intensity of the stress applied and the resulting strain hardness.

Hadfield, who was first to introduce this steel nearly 60 years ago, stated (1)[1] that its usual hardness in the as-quenched, purely austenitic state was equal to about 200 Brinell, and that by a suitable cold working the hardness could be increased to 550-580 Brinell; even slightly deformed steel was strain hardened appreciably.

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[1]The figures appearing in parentheses refer to the bibliography appended to this paper.

Hopkinson and Hadfield (2) offered a diagram reproduced in Fig. 1 showing the strain hardenability of austenitic manganese steel, high nickel steel (22.8 per cent nickel), and mild steel as affected by predetermined degree of elongation caused by stretching the standard tensile strength specimens. While the original hardness of mild steel
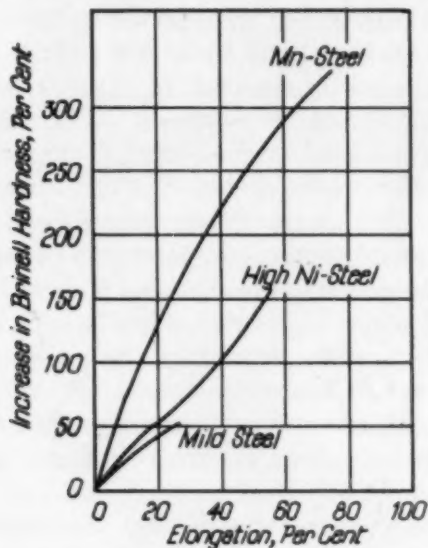

Fig. 1—Effect of Deformation on Strain Hardening of Manganese Steel, Nickel Steel, and Mild Steel Tensile Test Specimens. Original Hardness: Manganese Steel, 223 BHN., Nickel Steel, 164 BHN., Mild Steel, 137 BHN. *B. Hopkinson and R. Hadfield.*

was increased by some 50 per cent at the point of rupture, the hardness of the austenitic manganese steel rose to a value nearly 325 per cent greater than its original hardness of 223 Brinell. It is due to this tremendous increase in hardness under the conditions of cold work that manganese steel occupies its position of importance in the industry.

Krainer (3) likewise measured the increase in hardness of austenitic manganese steel tensile specimens 6 millimeters in diameter, tested to failure. He expressed such increase as a ratio between the ultimate and the original hardness values for steels of slightly different analysis:

Carbon	Manganese	Original Brinell Hardness	Hardening Ratio
1.10	13.60	211	2.26
1.20	13.61	236	1.98
1.31	14.18	217	2.55
0.97	13.58	226	2.09
0.92	14.45	205	2.33

——Analysis, Per Cent——

de Golyer (4) found that under cold work the hardness of manganese steel could be increased to 400 to 480 Brinell, and Evans (5) also estimated that cold working resulted in an increase in hardness of this steel from 200 to 400 Brinell. On the other hand, Hanks (6) claimed that by a suitable cold deformation of austenitic manganese steel its hardness could be increased to 700 to 800 Brinell in some classes of work. Fahrenwald (7) likewise advanced claims that under some conditions the hardness of manganese steel could be increased to about 700 Brinell by cold working. However, the general consensus of opinion among different investigators places the maximum strain hardness of manganese steel between 450 and 550 Brinell (see Hall (8), Evans and Burtt (9), Anon. (10), etc.).

Clark and Coutts (11) undertook quantitative measurement of the degree of strain hardening in manganese steels of slightly varying analyses. They cold-worked the specimens by subjecting them to uniform blows delivered by a hand operated hammer, for predetermined length of time. In this manner they obtained the following data:

Analysis of Steel Per Cent		Time of Hammering, Min.				
		0	15	30	50	65
Manganese	Carbon	Brinell Hardness				
12.37	1.31	202	247	444	477	532
12.07	1.15	187	235	418	495	555
10.70	1.12	179	228	387	477	512
10.27	0.97	217	247	444	512	578

This table illustrates clearly the gradual increase in the surface hardness of manganese steel specimens under continued hammering. It must be pointed out here that, while under these conditions the surface of manganese steel becomes very hard and highly resistant to abrasion, the underlying layers of the metal retain all the toughness of the original, undecomposed austenite and thereby excellently resist shocks.

Accordingly, if manganese steel is to possess the optimum wear resisting properties, it should be exposed to stresses of sufficient intensity necessary to develop the maximum strain hardness. However, under some conditions and in certain classes of service the attainment of maximum strain hardness within the metal proves to be a liability rather than an asset. Thus, Hanks (6) suggested this when he said that the hardening of manganese steel in service was also responsible for its weakness under reversing stresses.

Hall (8) stated that the fatigue limit of manganese steel was

comparatively low (39,000 pounds per square inch), only slightly less than its proportional limit. As stressing beyond the elastic limit naturally resulted in cold work and the strain hardening associated with it, this steel did not resist well repeated stresses exceeding its elastic limit. Such stresses resulted in a rapid hardening of the parts subjected to them, with consequent loss in toughness.

Caulle (12) offered from his own experience a very conclusive proof of the preceding observation: He encountered frequent breakage of manganese steel pins used in dredge bucket chain, occurring before the surface of the pins showed any signs of wear. The breakage was eliminated by making the pins stationary, so as to prevent reversal of stresses within them. To find a satisfactory explanation for this phenomenon, Caulle sectioned transversely both the turning and the stationary types of the pins, and measured the distribution of hardness along their diameter:

── Turning Pin ──		── Stationary Pin ──	
Distance from the Surface	Brinell Hardness	Distance from the Surface	Brinell Hardness
40 millimeters (center)	193	39 millimeters (center)	185
27	245	29	195
22	324	23	203
13	324	11	215
6	370	6	240
Outside	486	Outside	390

Thus, under the conditions of reversing stresses the strain hardening effect not only extended much deeper below the surface, but also resulted in a definitely higher surface hardness.

In the light of the foregoing data it appears that little information could be actually found in the literature on the subject of the extent of the strain hardness capable of taking place in austenitic manganese steel. In an investigation of this property of manganese steel carried out by the present writer, it was found necessary therefore to develop a method for quantitative measurement of the increase in hardness as well as of the depth of its penetration below the strained surface, taking place on cold working of manganese steel.

In the following pages is reported a part of this investigation dealing with strain hardenability of austenitic manganese steel of the usual composition, at room temperature.

EXPERIMENTAL PROCEDURE

Commercial manganese steel in the form of ⅜-inch hot-rolled

bars was used throughout these tests. The analysis of this steel was 12.5 per cent manganese, 1.10 per cent carbon, 0.35 per cent silicon, 0.029 per cent phosphorus and 0.030 per cent sulphur. Under the microscope it appeared to be fairly clean and free from slag inclusions, but contained occasional stringers extending in the direction of rolling, which etched dark by boiling alkaline sodium picrate; these were apparently due to the dendritic segregation of carbides taking place during the freezing stage.

Pieces of this material about 1 inch long were packed in an iron container and sealed to prevent their oxidation and scaling. They were heated for 30 minutes at 1920 degrees Fahr. (1050 degrees Cent.) and immediately quenched in cold water. This treatment resulted in purely austenitic structure, although the segregated carbides were not apparently completely eliminated. On reheating these specimens above the temperature at which precipitation of carbides begins, the stringers of these carbides invariably re-appeared.

The cold working step was performed under a specially constructed drop hammer consisting of a cast iron block weighing 43 pounds, sliding along a pair of vertical guides. The work input was controlled by dropping the weight from a predetermined height, the maximum capacity being about 350 foot-pounds. To maintain a flat surface in the part of the hammer impinging upon the specimens, a piece of 1-inch hardened tool steel was inserted in it in such way as to form hard working surface.

The test specimens were placed upon a cast iron anvil and held in place, in vertical position, by means of a yoke made of 1-inch round bar of steel, 0.5-inch high, with ⅜-inch hole drilled in the center. This yoke was provided with a radial slit which permitted its extension as the specimen was upset under the impact, thus avoiding jamming of the specimen in the yoke. For relative arrangement of the yoke and specimen see Fig. 2.

To attain critical stresses with a drop hammer of only 350 foot-pounds capacity, it became necessary to reduce the cross section of the specimens. This was done most readily by grinding one end of the latter to the form of a 60-degree cone, whereby a gradient of stresses was obtained in the conical section of the cold-worked specimen, varying from extremely high stress concentration at the very apex of the cone, to relatively low ones at the base of the cone and at the cylindrical part of the specimen.

Since manganese steel does not lend itself readily to machining,

grinding the specimens on an emery wheel was resorted to. The grinding operation was facilitated by constructing a lathe which permitted rotation of the specimen, mounted in a suitable chuck, at any desirable angle formed with the face of the grinding wheel, and also capable of linear motion in directions both parallel and perpendicular to the wheel axis.
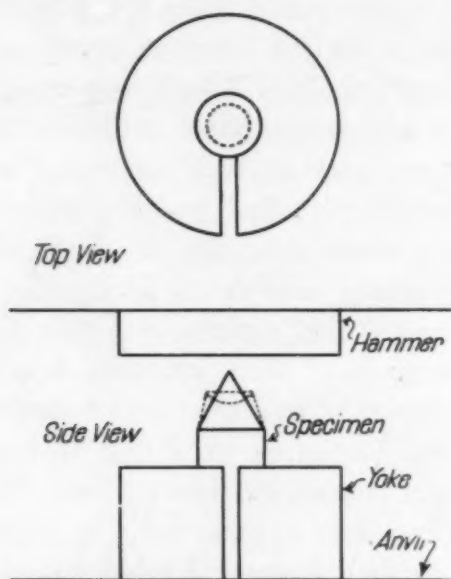

Fig. 2—Showing Relative Arrangement of Yoke and Specimen.

The experimental procedure thus consisted first of preparation of the test pieces from short lengths of ⅜-inch bars, which had been previously quenched to give wholly austenitic structure. These test pieces were then subjected to a specified number of impacts of desired intensity under the drop hammer. The hardness of the cold-worked surface thus produced was measured on the Rockwell C scale, with the specimen held vertically under the penetrator. To study progressive penetration of the strain hardness below the cold-worked surface, the specimens were next sectioned along their longitudinal axis. This operation was done again by grinding on an emery wheel, care having been taken to avoid excessive heating of the specimens. In this manner nearly one-half of each specimen was removed by grinding. The remaining half was mounted in bakelite to facilitate subsequent handling of the specimens both during their polishing for the microscopic examination as well as testing for their hardness.

The depth of penetration of the strain hardening effect was

measured by taking Rockwell A readings (diamond brale, under 60-kilogram load) along the longitudinal axis of the specimen, starting immediately below the cold-worked surface and proceeding away from it until the readings assumed the value corresponding to the original hardness of the specimen. The distance of each indentation below the cold-worked surface was measured by means of a micro-scale.

It was found in connection with these hardness measurements that the readings had to be taken on the Rockwell A scale in order to avoid elastic deformation of the bakelite mounting under excessive loads, as well as the danger of loosening the specimens in their mounting.

As the Rockwell A scale is relatively little used, it was deemed advisable to report all readings in terms of the commonly employed C scale. The necessary conversion was done by means of a table provided by the manufacturer of this testing machine. As this conversion table extends only to the value of C-20, and the values for the Rockwell A scale corresponding to lower readings are expressed in terms of the Brinell scale, it became necessary to co-ordinate these two scales and thus to eliminate the confusion which would have resulted from the use of two different scales in a single series of readings.

To check the reliability of such conversion procedure, the hardness of an unstrained specimen of manganese steel was measured first by taking five readings on the C scale and then the same number on the A scale; the same operation was repeated upon a specimen of manganese steel fully hardened by cold work:

Unstrained Specimen Rockwell		Cold-worked Specimen Rockwell	
C-8.1	A-55.5	C-51.2	A-75.7
8.3	54.5	50.9	76.1
10.4	55.1	49.8	75.8
7.2	56.8	50.6	74.9
8.0	54.9	49.3	76.5
C-8.4	A-55.3 or, C-9.6	C-50.4	A-75.8 or, C-49.8

Thus it was found in this particular case that the hardness conversion tables were accurate within one unit of the Rockwell C scale approximately.

It was realized from the start that the mounting of specimens in bakelite could have resulted in a certain source of error, because

this operation involved heating the specimens to 285 degrees Fahr. (140 degrees Cent.) and simultaneous application of a pressure as high as 2500 pounds per square inch. To ascertain the extent of this possibility, a blank test was made in which a specimen of quenched manganese steel was mounted in bakelite and its hardness determined on the Rockwell A scale. The average value of five readings was A-57.0, corresponding to C-12.6; on the other hand, the hardness of a specimen that had not been mounted was A-55.3 or C-9.6. Accordingly, there does appear to be some difference due to the manner of preparation of specimens. But this applies only to the range of low hardness values, which are solely of a general interest here. In the subsequent work, when the strain hardening effect was first measured directly by taking readings on the cold-worked surface, and then the depth of penetration of hardness was measured on the cross sectioned specimen mounted in bakelite, the surface hardness and the hardness of the immediately underlying layers of the specimen were found as a rule nearly equal.

Strain Hardening of Cylindrical Specimens

A number of cylindrically shaped specimens of manganese steel, about 15 centimeters long and 9.3 millimeters (⅜ inch) in diameter, was subjected to one impact of an intensity gradually increasing from 12.5 to 300 foot-pounds. The specimens were held vertically and the impact was delivered along their longitudinal axis.

The hardness of the cold-worked specimens was measured by taking five readings on the Rockwell C scale and determining their average value. Then the specimens were cross sectioned and the distribution of the strain hardness along the longitudinal axis was determined. In these tests it was unnecessary to resort to mounting the specimens in bakelite since the halved cylinders could be held easily on a V-shaped anvil under the penetrator of the hardness testing machine.

The distribution of hardness within these specimens, shown in Fig. 3, was far from being uniform. Contrary to all expectations, the maximum hardness values were not obtained at the cold-worked surface, but invariably were found some distance below it. Usually a fairly well defined gap separated the surface hardness from the corresponding maximum. The diagram shows this decline from the intermediate values at the surface down to minima located about

0.5 to 1.5 millimeter below the surface. Beyond this point the hard-
ness values increase rapidly, attaining their maxima at a distance
varying between 3 and 6 millimeters below the surface.

Since the strain hardening effect in austenitic manganese steel
is produced not so much by a static stress as by the actual flow of a
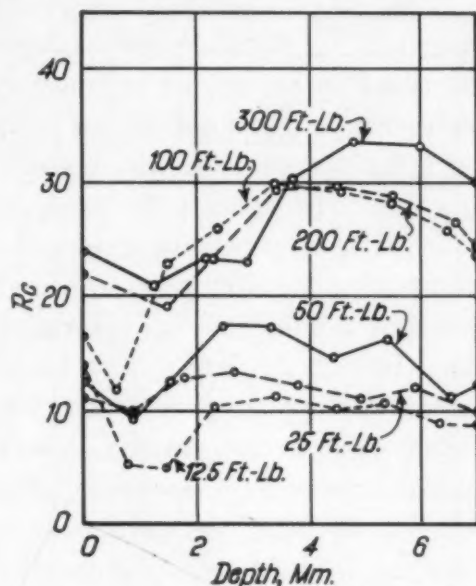

Fig. 3 — Depth of Penetration of
Strain Hardness in Cylindrical Specimens.

part of the metal in relation to the rest, the resulting hardness with-
in the specimen varies in proportion to relative amount of the de-
formation which has taken place under the impact. Accordingly,
the cold-worked surface proper does not by any means represent
the zone of maximum deformation—since it is merely pushed ahead
of the hammer into the soft metal underlying this surface austenitic
matrix. It is these underlying layers that are called upon to adjust
the resulting stresses, and it is therefore these layers that acquire
the maximum strain hardness. It stands to reason that the sides
of the cylindrical specimen are more or less pushed outward and
therefore do not undergo as much of kneading as does the central
portion of the specimen.

In general, it may be stated that under an impact of 300 foot-
pounds absolute, directed against a specimen ⅜ inch in diameter,
the maximum strain hardenability attained here was about Rockwell
C-34. This corresponds to a unit stress concentration of about
2500 foot-pounds per one square inch of the cold-worked surface.
And since, according to the literature, this steel could be hardened

to about 500 Brinell, or roughly C-50, a need was apparent for employing greater unit stress concentration. This was achieved by reducing the cross section of the specimens.

Strain Hardening of Cone-shaped Specimens

To permit attainment of ultimate strain hardness in austenitic manganese steel subjected to an impact of only moderate intensity, one end of the specimens was ground to the shape of a 60-degree cone. As a standard for comparison was chosen an impact of 50 foot-pounds intensity delivered against the apex of such cone representing the upper end of a vertically held specimen. It was found that under these conditions a pronounced strain hardening effect was obtained at the surface of a truncated cone formed under the impact (shown by dotted lines in Fig. 2), and that in the region immediately adjacent to this surface the specimen was somewhat upset. Due to the gradually increasing cross section in the direction away from the apex, the strain hardening effect decreased correspondingly until, at the base of the cone, it disappeared almost completely. With greater impact intensities, the hardening effect penetrated into the cylindrical portion of the specimen and there was also a tendency on the part of the latter towards bending.

It was found from the outset that no duplicate tests would give precisely the same data despite all the efforts taken to maintain the testing conditions identical. This was undoubtedly due to the lack of homogeneity in the specimens, such as difference in grain size, presence of zones of carbide enrichment, etc., and also to physical impossibility of delivering exactly the same blow in the same location, time after time. In several duplicate tests performed at the beginning of this series, it was occasionally found that a difference equal to two or three Rockwell C units existed between the corresponding readings of individual tests; however, both tests invariably yielded data of the same order. For this reason, it was decided to cover the field extensively rather than intensively: Both the intensity of impacts and their number were varied through a large range, using in most cases only a single specimen and hoping that under such conditions, even if the individual measurements were somewhat off from their absolute values, at least a general trend of behavior of austenitic manganese steel under conditions of cold work could be established.

Since it was desired to compare all other data with that obtained for a specimen strain hardened under one impact of 50 foot-pounds intensity, the latter test was performed in duplicate. In Fig. 4 the two sets of data are represented by differently marked series of dots, while the curve showing mean values is drawn equidistantly between these. It is apparent that a sizable difference
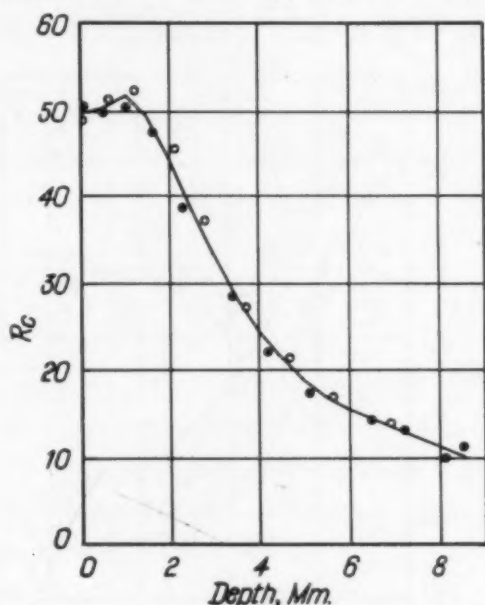

Fig. 4—Depth of Penetration of Strain Hardness in Austenitic Manganese Steel Specimen, Ground to 60-Degree Cone, Under One Impact of 50 Foot-Pounds (Absolute).

exists between these two sets of data, particularly in the region of high strain hardness.

Similarly to the case observed with the cylindrical specimens, the cold-worked surface does not represent in this diagram the conditions of maximum hardness. The peak of strain hardness, corresponding to the value of Rockwell C-52, was found about 1 millimeter below the surface; beyond this point the strain hardening effect began to decrease gradually until at a depth of about 8 millimeters below the surface, it practically disappeared and the hardness of the unstrained austenitic matrix remained unchanged.

The existence of the peak of maximum hardness probably indicates the occurrence of a center of greatest distortion of the metal a short distance below the cold-worked surface. This happened through inversion of the apex of the cone during the first stage of the deformation, and its forcing into the soft austenitic matrix as if the former were a dull tool. The half moon shown in Fig. 2

in dotted line underlying the cold-worked surface represents diagrammatically the outline of this "tool" and although this region bears the brunt of the impact, the shape of this portion of specimen is such as to prevent it from further deformation, and consequently it is forced into the underlying layers causing them to flow and thereby hardening them to a degree which is somewhat higher than its own hardness.
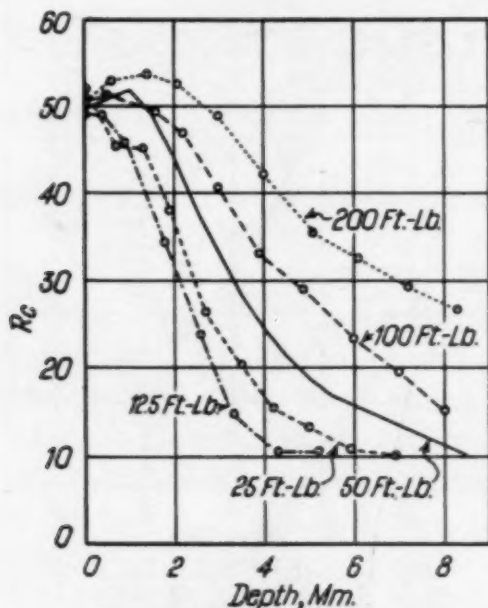

Fig. 5—Strain Hardening of Austenitic Manganese Steel Specimens, Ground to 60-Degree Cone, Under a Single Impact of Varying Intensity.

In this connection it is interesting to refer to the work of Burns, Moore, and Archer (13) on quantitative hardenability of steels during quenching. In their study of the depth of penetration of hardening the authors had established a characteristic curve which could be easily called a prototype of the curve depicted in Fig. 4 not only insofar as its general shape is concerned, but also on account of its peculiarities. One of such peculiarities observed by these investigators was the appearance of maximum hardness below the surface, the value of which was usually slightly above the surface hardness. As a possible explanation of this phenomenon they suggested retention of austenite at the surface, lower carbon content at the surface resulting from ingot segregation, or age hardening of martensite just below the surface. However, none of these hypotheses are valid in the present case. Neither could it be

explained by surface decarburization since the specimens were ground to shape after quenching.

When the series was expanded by varying the intensity of impact from 12.5 to 200 foot-pounds (Fig. 5), it was found that practically the same degree of hardness at the surface was obtained with the lightest as with the greatest impacts used. This could have been expected since with the conically-shaped specimens the application of a light impact gave at the apex of the cone exactly the same concentration of critical stress as with greater impacts—that is, infinity if we assume that the cross section area of the apex of a cone equals to zero. The chief difference consists largely of the degree of penetration of the strain hardening effect, since the effect resulting from the application of 12.5 foot-pounds could be expected to stop much sooner than the one taking place under 200 foot-pounds impact.

The curves representing the depth of penetration of strain hardness under the impacts of 12.5 and 25 foot-pounds respectively show no peak of maximum strain hardness characteristic of the other curves. Or rather, their peaks are probably so close to the cold-worked surface that their detection was impossible with the type of hardness measuring instrument used in the present investigation.

It appears that the greater the intensity of an impact, the deeper the peak of maximum hardness lies below the surface, and also the greater is the tendency to strain harden the cylindrical portion of the specimen.

Strain Hardening Under Repeated Impacts

In all the tests described so far, the strain hardness of manganese steel specimens was obtained under the conditions of a single impact of predetermined intensity. But since, under actual working conditions, the articles made of this steel are usually exposed to a great number of impacts rather than to a single one, it was pertinent to investigate progressive strain hardening under repeated impacts.

Fig. 6 represents the summary of a series of tests obtained on subjecting the cone-shaped specimens to repeated impacts of 12.5 foot-pounds intensity. With gradually increased number of the impacts, the hardness of the cold-worked surface was increased only

slightly; most of the energy delivered by the subsequent impacts was used, apparently, for increasing the hardness of the underlying layers.   However, it was found that both the surface hardness and the depth of penetration of hardness attained after four impacts were somewhat higher than those produced by only one impact; the same
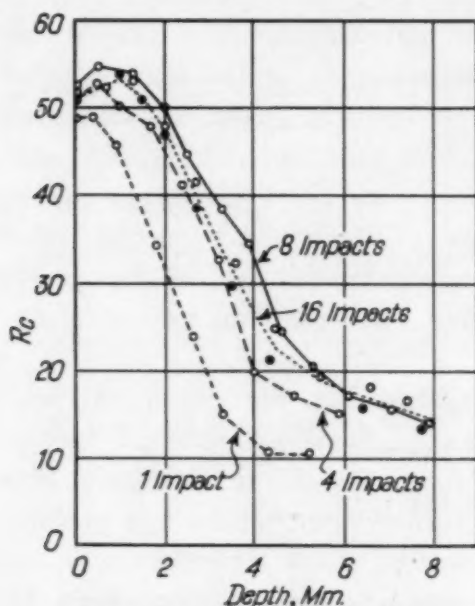

Fig. 6—Strain Hardening of Austenitic Manganese Steel Specimens, Ground to 60-Degree Cone, Under Repeated Impacts of 12.5 Foot-Pounds Intensity.

increment occurred when the number of impacts was increased from four to eight.   An apparent discrepancy was observed, however, when on plotting the data of a test corresponding to 16 impacts of 12.5 foot-pounds intensity it was noticed that the resulting curve was located inside the curve for the test involving eight impacts.   To confirm this inconsistency, the test was repeated—but again the new data, although slightly different from the previous set, was found to possess the same tendency.   These two different sets of data are indicated in Fig. 6 by differently marked circles and the curve corresponding to this number of impacts is drawn as the mean.

The same procedure was tried with impacts of 50 foot-pounds intensity and the results, indicated in Fig. 7, were practically of the same character as in the preceding test.   Here again increasing the number of the impacts to two resulted in a slightly greater surface hardness as well as deeper penetration of hardness below the cold-

worked surface; but when the number of impacts was increased to four, the hardness of the surface and the layers immediately under-lying it, actually decreased. The depth of penetration increased still further.
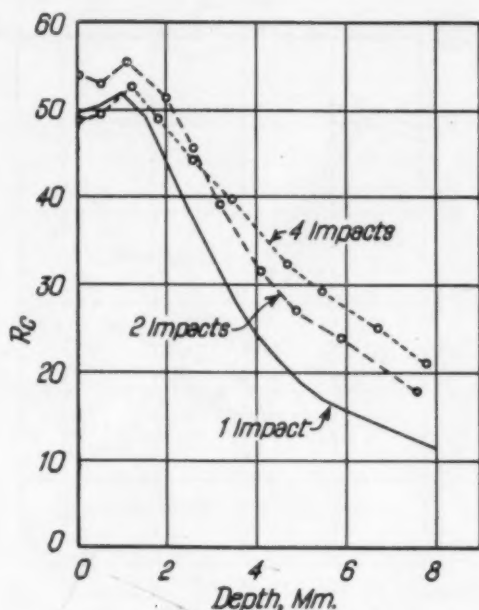

Fig. 7—Strain Hardening of Austeni-
tic Manganese Steel Specimens, Ground
to 60-Degree Cone, Under Repeated Im-
pacts of 50 Foot-Pounds Intensity.

This unexpected retrogression in the values of the maximum strain hardness of austenitic manganese steel at the surface is prob-ably due to the attainment of a point of satiation at which the metal underwent all the deformation it possibly could accommodate with-out rupturing. Any further deformation beyond this point actually resulted in an overstrained and weakened metal. Under the micro-scope, this condition was characterized by a tendency towards chipping and flaking at the cold-worked surface of the specimen. The hardened layers of manganese steel specimen became incapable of further absorbing the energy delivered by the additional blows and the latter was transferred to the layers located farther below the surface.

DETERMINATION OF CRITICAL UNIT STRESS

As it was mentioned before, in all tests involving the use of cone-shaped specimens a certain critical value for the unit concentration of impact was reached no matter how small was the absolute impact

intensity. Impacts of greater intensities resulted merely in greater depth of penetration without affecting appreciably the surface hardness. To find this critical concentration of energy per unit surface of manganese steel, required to develop the maximum strain hard-
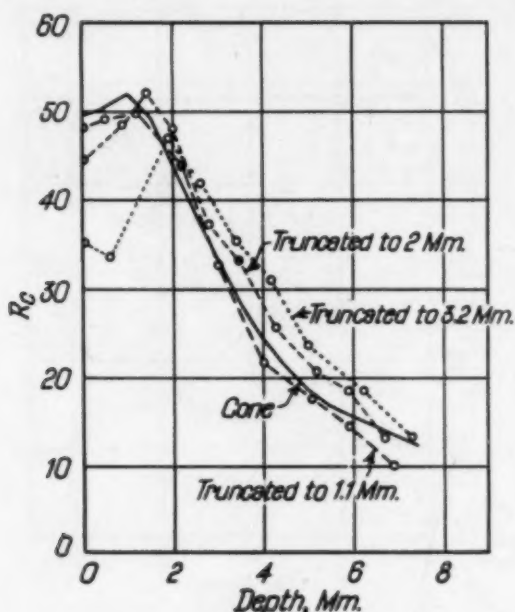

Fig. 8—Strain Hardening of Austenitic Manganese Steel Specimens, Ground to 60-Degree Truncated Cones, Under One Impact of 50 Foot-Pounds Intensity.

ness at the surface by a single blow, a brief series of tests was undertaken in which the specimens were prepared in the form of truncated cones having varying diameters of their upper limiting surfaces. This arrangement permitted quantitative estimation of the intensity of impact per unit area of the original surface.

Fig. 8 illustrates the results obtained in this series. The strain hardening effect obtained in the specimens having the diameter of their upper limiting surfaces equal to 2.0 and 3.2 millimeters respectively, was essentially the same as that obtained in the cylindrical specimens reported in the earlier part of this paper (see Fig. 3); if there was no gap indicated between the values for the surface hardness and the maximum hardness peak, it was likely due to the fact that in this case the hardness measurements were not carried close enough to the edge of the specimen.

In the specimen having the upper limiting surface equal to only 1.1 millimeter in diameter the distribution of the strain hardness approached rather closely the pattern found in the conically-shaped

specimen under the same conditions. In these two cases not only the hardness of the cold-worked surfaces was nearly the same, but also the depth of penetration of this effect was practically identical.

It would be of interest to calculate here the impact concentrations per unit area of the cold-worked surfaces employed in this series:

Diameter of the Upper Surface Millimeter	Area of the Upper Surface Square In.	Impact Concentration in Ft.-Lb. per Sq. In.	Surface Hardness Rockwell C
1.1	0.0014	35,700	48.2
2.0	0.0046	10,900	44.5
3.2	0.0116	4,300	35.0

Thus, when austenitic manganese steel is exposed to a single impact of a unit concentration equal to about 35,000 foot-pounds per one square inch, the steel is hardened almost to its ultimate strain hardness value.

This factor accounts for rather common failures of manganese steel in service when the conditions are such as to limit the intensity of cold work to low values. The fact is not always appreciated that, in order to get optimum properties from a wear resisting manganese steel article, it should be exposed originally to fairly large strains. Of course, it is not necessary to employ such large outputs of work as indicated above. It is quite conceivable to attain such critical strain under a blow of ordinary machinist's hammer delivered against a sharply pointed tool. This is undoubtedly what takes place when one attempts to cut manganese steel with a cold chisel and finds that, after an easily made first notch, most of the subsequent effort results in damaging the edge of the chisel. A small quantity of work applied to a limited zone of a strain hardening material is apparently just as effective in attaining the maximum strain hardness value as an impact of 35,000 foot pounds delivered against a piece measuring one square inch in cross section.

SUMMARY

1. A method for quantitative measurement of the strain hardening effect in austenitic manganese steel was developed. This consisted of delivering a single or repeated impact against a vertically held specimen, measuring the hardness of the cold-worked surface, and then cross sectioning the specimen and measuring the depth of

penetration of this effect, in the direction away from the surface.

2. Using cylindrical specimens, it was found that the hardness below the surface first decreased slightly and then began to increase, reaching a peak of the maximum value which appreciably exceeded the surface hardness.

3. Using conical specimens, the hardness of the cold-worked surface was found to be near its ultimate value disregarding the intensity of the impact. These cone-shaped specimens also possessed below the surface a peak of the maximum hardness which, however, was less pronounced. In the lower part of the specimen, the hardness effect gradually declined from this maximum value until the hardness of the original, unstressed metal was reached.

4. Under repeated impacts, a point invariably was reached when the resulting strain hardness immediately below the surface not only ceased to increase on further hammering, but actually decreased slightly. A greater number of light impacts was required to bring about this point of hardness satiation than was the case with more intense impacts. Once the outer layers of manganese steel became fully hardened, the strain hardening effect had a greater tendency towards penetration within the metal.

5. Rather unexpectedly high values were found to be required for strain hardening flat surfaces to the maximum value under the conditions of a single impact: To attain under these conditions the hardness of Rockwell C-50, the unit stress concentration was found to be as high as 35,000 foot-pounds per one square inch.

Acknowledgment

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DISCUSSION

Written Discussion: By Walter Crafts, research metallurgist, Union Carbide and Carbon Research Laboratories, Inc., Niagara Falls, N. Y.

The author has developed an ingenious means of determining maximum cold-worked hardness of Hadfield manganese steel. In our work, the hardness after straining has been determined by surface hardness tests after repeated hammering, by repeatedly dropping the scleroscope hammer, or by taking a Rockwell "C" or "A" determination in the base of a Brinell impression. The first two methods have given unsatisfactory results as the hardness seems to periodically drop off and return to a maximum. The Rockwell hardness at the base of a Brinell impression is not a maximum but is pretty consistent and varies for austenitic materials with respect to initial hardness as shown in Fig. 1.

Presumably a higher initial and, therefore, cold-worked hardness is desirable for wear resistance. This may be produced in austenitic manganese steel by changes of composition and heat treatment. Higher initial hardness, however, is almost invariably accompanied by brittleness to such a degree that it seems impractical to attempt to utilize the probable increase of wear resistance resulting from higher hardness. Furthermore, experience has been that failures usually result from deformation or cracking of metal back of the surface rather than by abrasion of the surface.

As indicated in the author's analysis of the literature, the greatest improvement in manganese steel can probably be realized by raising the yield point, as in the chromium-copper-manganese steel. The steels described below were forged to 1 inch in diameter bars and water-quenched.

Per Cent Carbon	Per Cent Manganese	Per Cent Chromium	Per Cent Copper
1.0	12.0	—	—
1.0	12.0	5.0	1.0

Yield Point pounds per square inch	Tensile Strength pounds per square inch	Elongation in 2 inches—Per Cent	Reduction of Area—Per Cent
56,000	139,000	43.0	43.1
75,500	140,500	38.0	34.1

It will be noted that the yield point is raised without changing the tensile strength or seriously harming ductility. Cylinders 2 inches high and 2 inches in diameter of similar cast steels were subjected to 12,600 blows of 1056 foot-
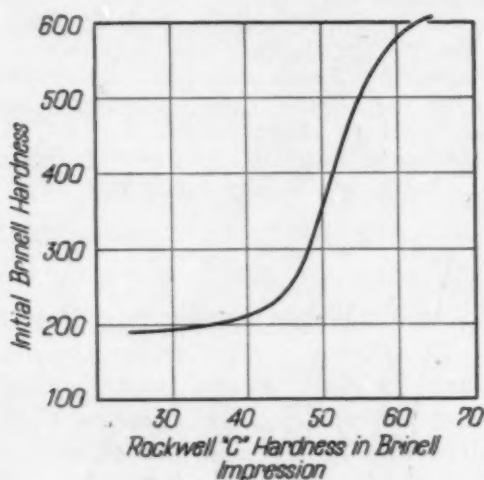

Fig. 1.

pounds under a Nazel forging hammer. The average loss of height of the ordinary Hadfield steel was ⅜-inch compared to only 1/16-inch for the 5 per cent chromium steel. The hardness of all hammered surfaces was 387 Brinell. Experience has shown that the improvement due to chromium is realized to an even greater degree in Hascrome, a hard facing austenitic steel containing even more chromium. The cold-worked hardness of this alloy is no greater than that of ordinary Hadfield steel, but it does have abnormally high wear resistance, in addition to the stiffer backing, probably as a result of the specific effect of chromium in resisting abrasion.

Author's Reply

The author wishes to thank Mr. Crafts for his most interesting discussion. His method for determination of the strain hardenability of austenitic materials could hardly be exceeded in simplicity; although, as it was pointed out, the results do not necessarily indicate the maximum strain hardness attainable. This deviation from the ultimate value becomes of a considerable magnitude in the case of austenitic manganese steel considered here: Since the initial hardness of this steel is only 200 BHN, it follows from the conversion curve

given in the discussion that the corresponding Rockwell C hardness at the bottom of Brinell impression should be in the neighborhood of C-30. This value is apparently far below the maximum hardenability of austenitic manganese steel as reported in the present paper.

Similarly to Mr. Crafts, the author also has tried determination of strain hardness of manganese steel after extensive hammering. These tests were carried out with the aid of a small, home-made drop hammer capable of delivering impacts of 5 to 10 foot-pounds intensity. Unit concentration of impacts as high as 1440 foot-pounds per square inch were attained by suitably reducing the hammered surface of the specimens made in the shape of truncated cones.

The results of these tests are summarized in the appended diagram. The curve showing surface hardness attained under the application of 360 foot-pounds per square inch is of a particular interest since it corresponds rather closely with the impact intensity used by Mr. Crafts in his work (1056 foot-pounds delivered against a 2-inch round face, or 336 foot-pounds per square inch). The author is happy to point out that the value of C-38.2 attained after 10,000 impacts delivered under these conditions is in a reasonably good agreement with the value of 387 BHN (Rockwell C-41) reported in the discussion.
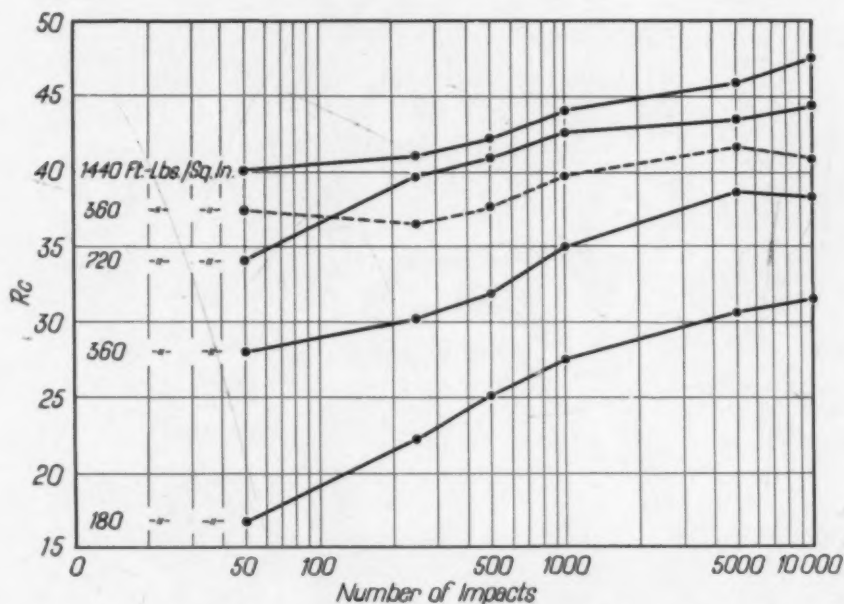
Surface Hardness of Austenitic Manganese Steel Specimens Cold-Worked by Various Number of Impacts of Indicated Intensities.

In regard to the relative merits of either high or low elastic limit values for austenitic manganese steel, the author wishes to draw attention to the 360 foot-pounds per square inch curve appearing in broken lines in the same diagram. It was obtained under identical test conditions, using a modified manganese-nickel steel (13.4 per cent manganese, 3.1 per cent nickel and 0.75 per cent carbon). This type of steel admittedly possesses a lower elastic limit and deforms more rapidly than the plain manganese steel. Although its ultimate strain hardness is only C-47 as compared with C-51 for the latter, it is ap-

parently capable of hardening under relatively light impacts much sooner than the ordinary manganese steel subjected to a similar degree of cold work. While this manganese-nickel steel would probably prove to be a failure when used in wear-resisting parts under impacts of great intensities, as in rock-crushers, there is a possibility that it would be superior to the plain austenitic manganese steel in places where it is exposed to only moderate impacts, such as are encountered in sand-handling machinery.